

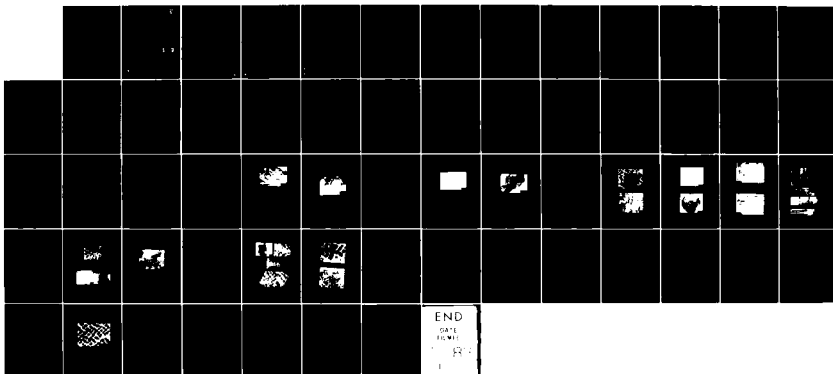
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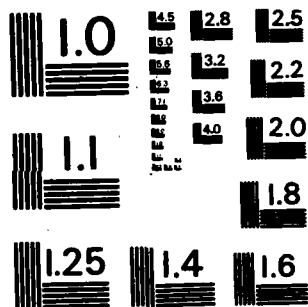
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20. ABSTRACT (Continue on reverse side if necessary and identify by block number) <b>High temperature low cycle fatigue (LCF) has been studied for directionally solidified (DS) and conventionally cast (CC) Rene' 80. For the conventionally cast material testing was carried out on smooth bars over the temperature range 750F (240C) to 1800F (982C). It was found that at low temperatures slip was planar and carbides acted as crack initiation sites. In correspondence with the importance of carbides on the fatigue life it was found that the life correlated best with total strain range (as opposed to plastic strain range)</b>		

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As the temperature increased, the life decreased and reached a minimum at 1400°F (760°C) corresponding to the maximum strength of  $\gamma'$  and the onset of  $\gamma'$  coarsening. Above this temperature the life increased and was determined by a trade-off between increased ductility (coarsening) and environmentally induced embrittlement. At 1800°F (982°C) the life was higher than that observed at room temperature. Tests were also done in which fatigue was carried out at 75°F to the half life; the temperature was then increased to 1400°F and fatiguing was continued. It was found that the low temperature damage had no effect on the fatigue life at 1400°F (760°C) and the life was the same as it would have been at 1400°F (760°C) had there been no prior fatigue at room temperature. Reversing the process had a profound effect. When the specimen was first cycled at 1400°F (760°C) and subsequently cycled at room temperature the residual life was essentially nil. This experiment showed that damage by deformation debris is reversible while environmental damage induced at 1400°F (760°C) is retained.

For the non-isotropic materials, most emphasis was placed on the DS alloys, a version of which is being used in the power plant of the B-1B bomber. For the standard composition of Rene' 80, it was found that for the same equivalent plastic strain range longitudinal specimens ([001] parallel to load axis) showed the longest life followed by 45° specimens ([001] at 45° to the load axis). The lowest life was exhibited by transverse specimens ([001] at 90° to the load axis). For a given orientation it was also found that the life depended on the orientation within the ingot and was a minimum for specimens taken from the center of the DS ingot. This was due to the low modulus which leads to a high total strain range and the occurrence of a script-like carbide in the interdendritic regions of this alloy.

When the DS material was tested as a function of temperature for a constant plastic strain range it was again found that the life was a minimum around 1400°F (760°C) and this was interpreted as being due to a maximum  $\gamma'$  strength which would accentuate environmental effects and structural stability. Above this temperature the life increased and this was interpreted as being due to  $\gamma'$  coarsening which leads to a reduced stress for a given plastic strain range.

A large dispersion in the data was observed for 45° specimens. At first this was thought to be due to specimen inhomogeneity or improper test techniques. However, a geometric analysis showed that for small misorientations from 45° there are large variations in both modulus and Schmidt factor, both of which have a profound influence on the fatigue life. It can thus be concluded that dispersion of the data for 45° DS specimens is characteristic of this orientation.

Attempts were made to modify the carbide morphology by Hf additions and to thereby increase the fatigue life. Although the Hf additions did produce a more benign carbide morphology (i.e., blocky as opposed to script-like), the life actually decreased when compared on the basis of constant plastic strain for all temperatures except 1400°F (760°C) where the life was equivalent to that for the conventional composition. The reasons for this were associated with the fact that the alloy was stronger (probably due to increased partitioning of the Ti to the  $\gamma'$ ) and crack propagation occurred more easily.

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mechanisms can interact to give lives between the two extremes. For example, a slip band can form in the vicinity of a carbide and due to the stress intensification of the slip band, crack the carbide. If carbide cracking is the controlling event during low temperature LCF, then the life would be expected to correlate better with the total strain than with the plastic strain, since the total strain is a measure of the displacement to which carbides would be subjected. This is confirmed in Fig. 5, where for 75°F the degree of correlation is significantly improved when the data are represented in terms of total strain, indicative of a critical displacement criterion for cracking. At 800°F and 1400°F both total strain range and plastic strain range were found to provide equally precise representations of the data.

Additional metallography was carried out to further characterize the crack formation process at 75°F and an excellent example of carbide cracking can be seen in Fig. 6. It is clear that carbide cracking and not decohesion is the operative process and the carbide, not the carbide/matrix interface is the weak link.

As the temperature increases, crack initiation becomes mainly intergranular and always occurs in a region where there are numerous carbides. Such a region is seen in Fig. 7 taken from a specimen tested at 1400°F. This would tend to indicate an increased environmental component to fracture at this temperature. Crack propagation was mainly transgranular at 1400°F indicating a different effect of the environment for the propagation stage.

The effect of temperature on the fatigue life is summarized in Fig. 8 where average values of the fatigue life at constant strain

for high and low rates are plotted as a function of temperature. The effect of increased temperature is initially to decrease the life. Since there was no significant difference in the deformation mode between room temperature and 800°F (and very little difference up to 1400°F) the decreased life is consistent with an increasing environmental component. At 1400°F the observed life is a minimum. This is in reasonable agreement with recent observations for pure Ni [3] where the effect of environment on reducing the creep life was a maximum near this temperature. Above this temperature we have previously shown that the  $\gamma'$  coarsens [1,2,4] and there is a trade-off between an increased environmental effect and increased ductility. The trade-off is such that increased ductility becomes the dominant factor and the life increases in the range 1400-1800°F.

Two specimens were tested at the same plastic strain range ( $\Delta\epsilon_p = 0.05\%$ ) at room temperature to half of their projected lives as determined by the Coffin-Manson curve. The temperature was then increased to 1400°F and testing was continued to failure. These tests were done at a strain rate of 0.5%/min. The residual lives of these two specimens were identical to the total life obtained previously at 1400°F. Apparently the damage mechanism at 1400°F is much more severe than that at room temperature. This point was reinforced by reversing the temperature sequence. When this was done, failure occurred during the initial loading cycles at room temperature.

This behavior is a dramatic demonstration of the path dependence of the damage process. For the former experiments, damage accumulation at room temperature was probably in the form of dislocation debris which was annealed out at high temperature leaving

only the high temperature damage process to operate. For the later experiments, oxidation or oxygen penetration and subsequent embrittlement could not be annealed out and the high stresses that are developed at room temperature in coordination with the oxidation led to premature failure.

### III. Standard Composition DS Rene' 80

#### A. Heat Treatment

The composition of the standard composition Rene' 80 is shown in Table I. Ingots were directionally solidified at a rate of 10 in/hr under a gradient of 250°F/in at the interface.

Small gage blanks (0.27 in. diameter x 1.75 in. long) were cut from the ingots (after discarding the top and bottom) in longitudinal, transverse and 45° orientations. The gage blanks were then brazed to end pieces and heat treated as follows:

1. Solution the as-machined DS blanks at 2200°F for 2 hours in a vacuum, followed by a helium gas quench.
2. The DS blank is then electroplated with a 0.0005 in. thick Ni layer.
3. The DS blank and IN 718 end plugs are then brazed using GE braze alloy B-28 in a vacuum. Brazing is done at 2200°F for 15 minutes followed by cooling to 2000°F in 6-10 minutes. If the braze joint is unsatisfactory, it is rebrazed.
4. Hold at 1925°F for 4 hours to simulate the coating cycle.
5. Age at 1550°F for 16 hours, and furnace cool to RT.
6. The IN 718 is then aged at 1325°F for 8 hours, furnace cooled to 1150°F at a rate of 100°F/hr, and held at 1150°F for 8 hours and then furnace cooled to RT. (All done in a vacuum.)



Details of the specimen preparation technique are given in [5]. The heat treatment described above develops the duplex  $\gamma'$  structure shown in Fig. 9. After brazing, standard LCF specimens were final machined, the last step of which involved low stress grinding to minimize machined-induced residual stresses.

## B. Metallography of Initial Structure

### 1. As Heat Treated Rene'80

Typical carbide and grain boundary configurations are shown in Figs. 10 and 11.

The grain and carbide structure of this alloy depends on the distance from the bottom (chilled) end of the ingot and position dependent structural differences are shown in Fig. 12.

### 2. Grain Size Determination

Grain size measurements were done on a transverse cross-section of the 1-5/8 in. dia. x 6 in. long bar (#489) at the bottom, middle and top. The bottom end of the bar was cut approximately 2 to 3 inches from the chilled end of the original ingot.

The approximate grain sizes were:

- a) Bottom: 1 to 1.5 mm avg. dia.
- b) Middle: 2 to 4 mm avg. dia.
- c) Top: 3 to 8 mm avg. dia.

In the longitudinal sections the grains were usually longer than the gage section of the fatigue specimens (i.e., >0.75 in.). Micrographs of the different sections of the as-received Rene' 80 bar are shown in Fig. 12 while Fig. 10 shows typical grain sizes in LCF specimens taken from the bottom and top of an ingot.

### 3. Carbide Size and Morphology

Near the bottom of the ingot where cooling was rapid, carbides were mostly plate-like and crystallographic such as those seen in Fig. 11a. Near the top they were less crystallographic, the fraction of equiaxed carbides increased, and they were less concentrated on the grain boundaries. EDAX analysis of the carbides showed a very high Ti peak characteristic of MC.

### C. Mechanical Testing

The LCF behavior of DS Rene' 80 was primarily studied in the temperature range 75-1600°F. All tests were carried out using cylindrical specimens in the total strain control mode with periodic adjustments to maintain the plastic strain range constant. The cycle character was nominally zero-tension-zero ( $A_{\epsilon} = \Delta\epsilon/\bar{\epsilon} \approx 0.95$ ).

The gage sections of the smooth bar specimens were machined from DS ingots in the longitudinal, transverse and 45° orientations. Specimens were identified as to their location (top, middle and bottom) in the DS bar such that each specimen could be placed in a definite position in the original ingot. After testing, emphasis was placed on analysis and microscopy studies as described below.

### D. Results and Discussion

#### 1. Effect of Temperature on LCF Behavior

The LCF behavior for transverse and 45° specimens was studied over the temperature range 75-1600°F at a plastic strain range of 0.1%. The results are shown in Fig. 13. It can be seen that the life increases up to about 1400°F after which it increases at 1600°F to a level higher than that observed at room temperature.

Specimens were subsequently metallographically examined and the following observations were made:

a) Up to 1400°F, well-developed slip bands were present as seen in Fig. 14

b) Specimens tested at 75°F and 800°F showed a higher dislocation density in the slip bands compared to 1400°F.

c) Precipitate shearing by the dislocations was observed at 75°F as seen in Fig. 15

d) At 1400°F slip band broadening was observed implying a larger component of thermally activated cross slip than was seen at the lower temperatures. This conclusion is borne out by the tensile data shown in Fig. 16 where it can be seen that the yield stress starts to drop off rapidly after 1400°F. Such behavior is expected since the  $\gamma'$  precipitate has a maximum strength in the temperature range 1450-1550°F. Dislocation substructures at 1400°F are seen in Fig. 17.

e) For those specimens tested at 1600°F, no planar slip bands were observed and the  $\gamma'$  precipitates had significantly coarsened, showing interfacial dislocations at the  $\gamma/\gamma'$  interface similar to observations made for conventionally cast Rene' 80 [1,2]. The dislocation substructure is shown in Fig. 18.

The experimental observations are compatible with the idea that the life is determined by different mechanisms at different temperatures. At 75°F the life is determined by an accumulation of dislocation debris and carbide cracking. As the temperature increases the environment becomes increasingly important and the life decreases until at some point, around 1600°F, the  $\gamma'$  pre-

cipitates begin to coarsen. Above 1600°F the stress in a strain controlled cycle decreases rendering embrittlement by oxidation less effective in causing failure. In other words the trade-off between environmental damage and structural coarsening (which is beneficial) is biased in favor of structural coarsening at high temperature.

Such behavior is consistent with an approach developed previously under the auspices of this program in which crack initiation occurs when the following equation is obeyed [1,2]:

$$\sigma_{\max}^i \cdot l_i^P = C_o \quad (1)$$

where  $\sigma_{\max}^i$  = maximum stress in fatigue cycle at initiation

$l_i$  = relative depth of oxygen prepenetration

P,  $C_o$  = material constants

Experimental data was used to check the validity of this approach for all specimens tested at 1600°F and the results are shown in Figs. 19-21. It can be seen that the data correlates reasonably well to Eq. 1 and that there is a further improvement when the correlation is made in terms of  $\Delta\sigma$  rather than  $\sigma_{\max}$ . It is also noteworthy that the best correlation was obtained for the 45° specimens where stresses were the highest and where boundaries intersect the specimen surface, the conditions for which the model is expected to work best.

At 1600°F additional testing was carried out for plastic strain ranges from 0.02% to 0.25% and the results are shown in Fig. 22 of Coffin-Manson plots in Figs. 22-24. For transverse specimens, Fig. 22, the life depends on location in the ingot, at least for the highest plastic strain range (0.25%). The lowest

life was for specimens taken from the middle while the longest life was associated with specimens taken from the bottom. A similar trend was observed for the longitudinal specimens at all strain ranges, Fig. 23. An insufficient number of 45° tests were run to draw any conclusions as to the effect of location. The dependence of life on location, at least for the longitudinal specimens can be explained on the basis of modulus and a tendency to form a script-like carbide. The modulus as a function of location is shown in Fig. 25 by the set of graphs on the left side. The modulus decreases from bottom to top for the longitudinal specimens. This can be understood in terms of the average orientation becoming closer to [001] near the top. The [001] direction is, of course, "soft" and for a given plastic strain range, the elastic component will be greater the closer the [001] direction is to the stress axis. Thus near the middle, the total strain is large for a given plastic strain range (as seen on the graph on the right) and the tendency to form script-like carbides is also high. These two factors (high total strain and undesirable structure) act so as to minimize the life. On the other hand, near the bottom the modulus is high and the elastic strain component will be low. In addition, there is little tendency to form script-like carbides at this location. Both of these factors act in such a way as to increase the life. The effect of modulus and carbide structure on the LCF life is summarized in Table II.

The results for all orientations of the DS alloy are compared to those for the CC alloy in Fig. 26. At low lives and high

strains the differences are minimal. However, with decreased strain the lives for all orientations of the DS will exceed those that were observed for the CC material. Such a result is, of course, useful for applications since the low strain-long life regime is precisely within the operational envelope.

#### IV. Hf Modified Rene' 80

##### A. The Role of Hf

The research done on both the CC Rene' 80 and the DS Rene' 80 having the same nominal compositions as the CC alloy has shown that grain boundaries are frequently the sites for crack initiation. Such crack initiation is significantly affected by impurities that segregate to these regions such as O and S. Additions of Hf were made with the goal of improving the grain boundaries by occupying grain boundary sites and shunting S and O to the matrix where they are less harmful. In addition, Hf acts as a "getter" for these elements, tying them up as inclusions in the matrix where they should be relatively harmless. Hf has the additional effects of segregating to the  $\gamma'$  and causing a significant refinement in the  $\gamma/\gamma'$  structure. There also tends to be a refinement of the carbide structure which should result in an additional improvement in LCF performance. A micrograph comparing the Hf modified microstructure to that of conventional composition Rene' 80 is shown in Fig. 27.

##### B. Test Program

Hf modified bars were obtained from the General Electric Company and their composition is shown in Table I. These bars were used

to study effects such as temperature, specimen location in the ingot and specimen orientation on LCF behavior. The goal here was not to provide a complete characterization of LCF performance over a wide range of plastic strains as was done for the CC alloy and, to some extent, for the DS alloys. Rather, testing was done at a strain rate of 50%/min and at a plastic strain range of 0.1% over a range of temperatures. Metallographic studies similar to those described in previous sections were also carried out.

### C. Results and Discussion

The structure of the  $\gamma'$  is shown in the TEM micrograph of Fig. 28. It can be seen that the structure is similar to that for the conventional composition material shown in Fig. 9 with the exception that the heat treatment also appears to develop fewer small  $\gamma'$  particles.

The effects of orientation and temperature on LCF life are shown in Fig. 29 where the same general trends that were observed for the standard composition Rene' 80 were also observed for the Hf modified version. In particular, the life decreased with increasing temperature and depending on the orientation reached a minimum in the temperature range 1000°F-1400°F although the variation of life with temperature was minimal for the transverse specimens up to 1600°F. Comparison of Figs. 29 and 13 shows that at least for the Hf alloys used in this program the life was not improved in comparison to the standard composition except for transverse specimens at 1200°F and 1400°F and 45° specimens at 1400°F. Reasons for this may lie in the fact that the Hf modified version was stronger than the conventional alloys and even though the carbide structure

was more benign (e.g., Fig. 27), the increased strength offset any such potential advantage. The strength effect is shown in Figs. 30 and 31 where the maximum stress in a hysteresis loop ( $\Delta\epsilon_p = 0.1\%$ ) is plotted for both Hf modified and standard composition Rene' 80. Comparison of these figures shows that the Hf modified alloy is significantly stronger at 1600°F for all orientations. Application of Eq. 1 implies a shorter critical length of damaged zone and a somewhat lower fatigue life, which was actually observed.

In all cases and similar to what was observed for the standard composition, cracks formed where there was a high concentration of carbides, such as in interdendritic regions suggesting that improvements in life could be achieved by reducing the volume fraction of carbides.

For both the Hf modified and standard composition Rene' 80, specimens taken at 45° to the growth axis ([001]) showed significant dispersion in fatigue life. This can be explained on the basis of small misorientations from the nominal angular orientation. For longitudinal specimens a 10° misorientation from [001] causes an increase in Young's modulus of approximately 10% as shown in Fig. 32. The situation is completely different for the 45° specimens. A misorientation of 10° from a 45° angle between [001] and the tensile axis can lead to a difference in Young's modulus of anywhere from -18% to +10% and a variation of  $\pm 17\%$  in the Schmidt factor as shown in Fig. 33. Rotation of grains about the stress axis does not change the situation for longitudinally oriented grains, either in terms of Young's modulus or the Schmidt factor. However for 45° specimens there are very large variations. For example, consider two adjacent



grains in which the stress axis is [111] and [313], respectively. They make angles of  $54^\circ$  and  $47^\circ$ , respectively, with respect to [001] and may be brought into coincidence by a  $10^\circ$  tilt and a  $20^\circ$  rotation. Such differences were frequently observed as shown in Table III. The results quoted in Table III were obtained from in situ orientation studies done on thin foils on the transmission electron microscope. The Schmidt factor for [313] is 60% higher than that for [111] and is near the maximum difference that would be expected for two adjacent grains. Thus there could be a situation in which one grain may have a neighbor with a 60% higher yield stress and a 20% lower Young's modulus. Since the  $45^\circ$  orientation allows some independence in deformation of grains, significant differences in deformation behavior, strength and fatigue life are expected and this was observed.

## V. Summary and Conclusions

### A. Conventionally Cast Rene' 80

1. Crack initiation occurs on slip bands and at carbides at  $75^\circ\text{F}$ . Initiation at carbides causes a significant decrease in life.

2. The life decreases with increasing temperature up to  $1400^\circ\text{F}$ . Above  $1400^\circ\text{F}$  the life begins to increase and at  $1800^\circ\text{F}$  it exceeds the life at  $75^\circ\text{F}$  when correlated in terms of plastic strain range. The life is longer the slower the strain rate.

3. At  $1400^\circ\text{F}$  and above, environmental effects played a dominant role on the initiation and early propagation of fatigue cracks and boundary cracking became the dominant failure mode at  $1400^\circ\text{F}$  and above.

4. Testing at 75°F to 50% of the life and then continuing at 1400°F did not reduce the life compared to specimens tested only at 1400°F. This means that matrix damage was either recovered or did not interact with the grain boundary mechanism responsible for failure at this temperature. On the other hand, grain boundary damage in LCF specimens tested at 1400°F to 50% of the life reduced the remaining life at 75°F to one cycle (failure during initial loading).

B. Directionally Solidified Rene' 80

1. The life of DS Rene' 80 exceeds that of the CC material at 1600°F for all orientations investigated. The advantage of the DS material increases with decreasing strain range.

2. The life was generally a maximum for longitudinal specimens and a minimum for transverse specimens. Superimposed on that effect was a dependence of specimen location with respect to the original ingot for standard composition Rene' 80. The maximum life was observed for specimens near the bottom and the minimum was for specimens taken from the middle.

3. The low life of specimens taken from the central portion of the ingot was attributed to script-like carbide and low modulus as this reason. The low modulus gave rise to a large displacement which cracked the carbides.

4. The life for both the standard composition Rene' 80 and the Hf modified version decreased with increasing temperature up to about 1000-1400°F for all orientations and increased above 1400°F to levels in excess of the life at 75°F. This was similar to the observations for the CC alloys. This behavior

was attributed to a high strength up to 1400°F and an increased environmental effect. Above 1600°F  $\gamma'$  coarsening occurred and the stress dropped off leading to a longer life.

5. Addition of Hf changed the carbide structure from script-like to blocky but did not increase the life. In fact the life was, in some cases, lower than for the standard composition alloy. This was attributed to an increased strength (presumably a result of Hf changing the  $\gamma'$  character) of the Hf bearing alloy.

6. The dispersion of test results associated with 45° orientations could be understood in terms of modulus and Schmidt factor effects. For 45°, small misorientations give rise to large variations in both Schmidt factor and modulus. Since failure is a result of plasticity (Schmidt factor) and displacement (Young's modulus) these variations would be expected to cause a dispersion in the life.

## VI. Reporting Activity

### A. Presentations and Publications

#### 1. Presentations

- a. Evaluation of Damage Mechanisms at Intermediate Temperatures in Rene' 80. B. Bousier and S. D. Antolovich. AIME Fall Meeting, Louisville, KY, October 1981.
- b. High Temperature Fatigue Studies of Directionally Solidified Rene' 80. A. Prakash, K. McCurdy and S. D. Antolovich. AIME Fall Meeting, Louisville, KY, October 1981.
- c. Notch Low Cycle Fatigue Behavior of Rene' 80 at Elevated Temperature. P. A. Domas and S. D. Antolovich. AIME Fall Meeting, Louisville, KY, October 1981.
- d. The Effect of Microstructure on the Fatigue Behavior of Ni Base Superalloys, 27th Sagamore Conference, Bolton Landing, NY, July 1980.
- e. The Effect of Environment on the High Temperature Fatigue Behavior of Ni Base Alloys. Stephen D. Antolovich, invited paper presented at the 1982 Annual AIME meeting, Dallas, TX, February 1982.

#### 2. Publications

- a. Low Cycle Fatigue Behavior of Rene' 80 at Elevated Temperatures. Met. Trans., 12A, 1981, p. 473, S. D. Antolovich, S. Liu and R. Baur.

- b. The Effect of Microstructure on Fatigue Behavior on Ni Base Alloys, Fatigue: Environment and Temperature Effects, J. J. Burke and V. Weiss, eds., Plenum Press, NY, 1983, p. 119, S. D. Antolovich and N. Jayaraman.
  - c. Metallurgical Instabilities during the High Temperature Low Cycle Fatigue of Nickel Base Alloys, Mat. Sci. & Engr., 57, 1983, p. L9-L12, S. D. Antolovich and N. Jayaraman.
3. Publications in Preparation
- a. An Integrated Local Energy Density Approach to Notch Low Cycle Fatigue Life Prediction. P. A. Domas and Stephen D. Antolovich. Paper prepared and being internally reviewed before submission to journal. Anticipate submission to Engineering Fracture Mechanics.
  - b. A Mechanistically Based Model for High Temperature Notched LCF of Rene' 80. P. A. Domas and Stephen D. Antolovich. Paper prepared and being internally reviewed before submission to journal. Anticipate submission to Engineering Fracture Mechanics.
  - c. Effects of Environment and Structural Stability on the Low Cycle Fatigue Behavior of Ni Base Superalloys. Stephen D. Antolovich and N. Jayaraman. Paper prepared and being internally reviewed before submission to journal. Anticipate submission to High Temperature Technology.

- d. Microstructural Effects in the Fatigue Behavior of Directionally Solidified Rene' 80. Stephen D. Antolovich, A. Prakash, K. McCurdy and N. Jayaraman. Manuscript prepared and internal review complete. In process of revising before submission to journal. Anticipate submission to either High Temperature Technology or Metallurgical Transactions.
- e. The Effects of Temperature, Composition and Orientation on the Low Cycle Fatigue Behavior of Directionally Solidified Rene' 80. M. Raguet and S. D. Antolovich. First draft of paper completed. Anticipate submission to Journal of Materials Science.

**B. Dissertations and Theses****1. Ph.D.**

- a. Paul A. Domas. An Investigation of Notch Low Cycle Fatigue Life Behavior of Rene' 80 at High Temperature. Ph.D. dissertation. University of Cincinnati, June 4, 1981.
- b. A. Prakash. Low Cycle Fatigue Behavior of the Directionally Solidified Nickel Base Superalloy Rene' 80. Ph.D. dissertation. University of Cincinnati, December, 1981.

**2. M.S.**

- a. Bruno Boursier. Evaluation of Damage Mechanisms in the Ni Base Superalloy Rene 80 under Low Cycle Fatigue in the Temperature Range 75°F-1400°F. M.S. thesis. University of Cincinnati, December, 1981.
- b. Kevin J. McCurdy. The Effects of Temperature and Orientation on the Low Cycle Fatigue Behavior of DS Rene' 80 H. M.S. thesis. University of Cincinnati, expected January 1984.

**3. B.S.**

- a. J. Wukusick. Fatigue Crack Propagation of CC Rene' 80 in the Temperature Range 75°F-1400°F. Senior Thesis. University of Cincinnati, June 1982.

- b. Kevin J. McCurdy. Influence of Orientation and Environment on the LCF Life of Directionally Solidified Rene' 80. Senior thesis, University of Cincinnati, April 17, 1981.
- c. Daniel A. Wagner. Investigation of the LCF Life of Transverse DS Rene' 80 at Elevated Temperatures. Senior thesis. University of Cincinnati, June 1982.



VII. Participants in AFOSR 80-0065

<u>Name</u>	<u>Nature of Participation</u>	<u>Dates</u>
S. D. Antolovich	Principal Investigator	1/01/80- 6/30/83
B. Bousier	Research Engineer/MS Student	1/01/80 - 12/31/81
A. Prakash	PhD Student	1/01/80 - 11/30/81
K. McCurdy	BS Senior Thesis & MS Student	9/06/80 - 6/30/83
P. Domas	PhD Student	1/01/80 - 6/04/81
M. Raguet	Research Engineer	4/01/82 - 6/30/83
S. Liu	Post-doctoral Fellow	1/01/80 - 3/31/80
D. Wagner	BS Senior Thesis Student	9/06/80 - 6/15/81
K. Hemker	BS Student Laboratory Asst.	6/01/81 - 6/15/83
J. Wukusick	BS Senior Thesis Student	9/28/81 - 6/15/82
C. Wukusick	GE Employee-supplied DS Rene' 80	1/01/80 - 6/30/83
A. Rebillard	Student Assistant working with J. Wukusick	9/20/81 - 4/15/82

VIII. Interactions with Other Air Force Personnel and  
OSR Sponsored P.I.'s

The author has maintained contact with Drs. R. Pelloux and J. Weertman as well as with Drs. Reimann and Nicholas of the AFML. In addition, there has been technical interaction with Drs. Starke and Lankford on other aspects of fatigue problems.

References

1. Stephen D. Antolovich, S. Liu and R. Baur: Low Cycle Fatigue Behavior of Rene' 80 at Elevated Temperatures. Met. Trans., 12A, 1981, p. 473.
2. Stephen D. Antolovich, S. Liu and R. Baur: A Mechanistically Based Model for High Temperature LCF of Ni Base Superalloys Superalloys 1980, Eds. Tien, Wlodek, Gell and Maurer. ASM 1980, pp. 605-613.
3. R. H. Bricknell and D. A. Woodford: Met. Trans., 12A, 1981, p. 425.
4. Stephen D. Antolovich and P. Domas: High Temperature Fatigue of Rene' 80. Presented at International Fatigue Conference, Sherbrooke, Canada, July 1978.
5. A. Prakash: Ph.D. Dissertation. University of Cincinnati, December 1981.




Table I. Composition of Rene' 80 Alloys Used in this Study.

ALLOY	Composition (W/O)											
	C	Cr	Co	Mo	W	Ti	Al	B	Zr	Hf	Mg	Ni
Conventionally Cast R80	0.17	14.0	9.8	4.0	3.9	5.0	3.0	-	-	-	-	Bal
Standard (DS R 80)	0.16	13.8	9.2	4.0	4.0	4.9	3.04	0.015	0.02	0.05	0.07	Bal
Hf Modified Low Carbon (DS R 80H-LC)	0.073	12.9	9.6	4.0	4.9	4.48	3.02	0.015	0.01	0.74	-	Bal
Hf Modified Normal C (DS R 80 H)	0.15	13.72	9.39	3.91	4.10	4.84	3.04	0.015	0.01	0.72	6ppm	Bal

Table II. Effect of Modulus and Structure on LCF Life  
of Longitudinal and Transverse Specimens

LOCATION	MODULUS	$\Delta\epsilon_1$ (for a given $\Delta\epsilon_p$ )	TENDENCY TO FORM SCRIPT MC CARBIDE	RELATIVE LIFE (for a given orientation)
Top	Low	High	Intermediate	Intermediate
Middle	Intermediate	Intermediate	High	Low
Bottom	High	Low	Low	High

Table III. Misorientations Between Adjacent Grains.

Specimen Number	Number of Grains	Interdendritic Space (mm)	Average Misorientation from Stress Axis	Rotation 	Tilt 	Twist 
LT1	2	.25	--			
LT2	3	.23	5°	45°-40°-3°	2°	3°
LT3	3	.21	6°	20°-10°-35°	12°	
LT4	3	.27	7°	35°-35°-35°	8°	4°
LB1	3	.22	--	30°-15°-35°		
LB2	3	.21	--	35°-35°-30°		
LB3	7	.22	--	30°-35°-40°	4°	
LM1	3	.22	3°	45°-20°-25°	2°	4°

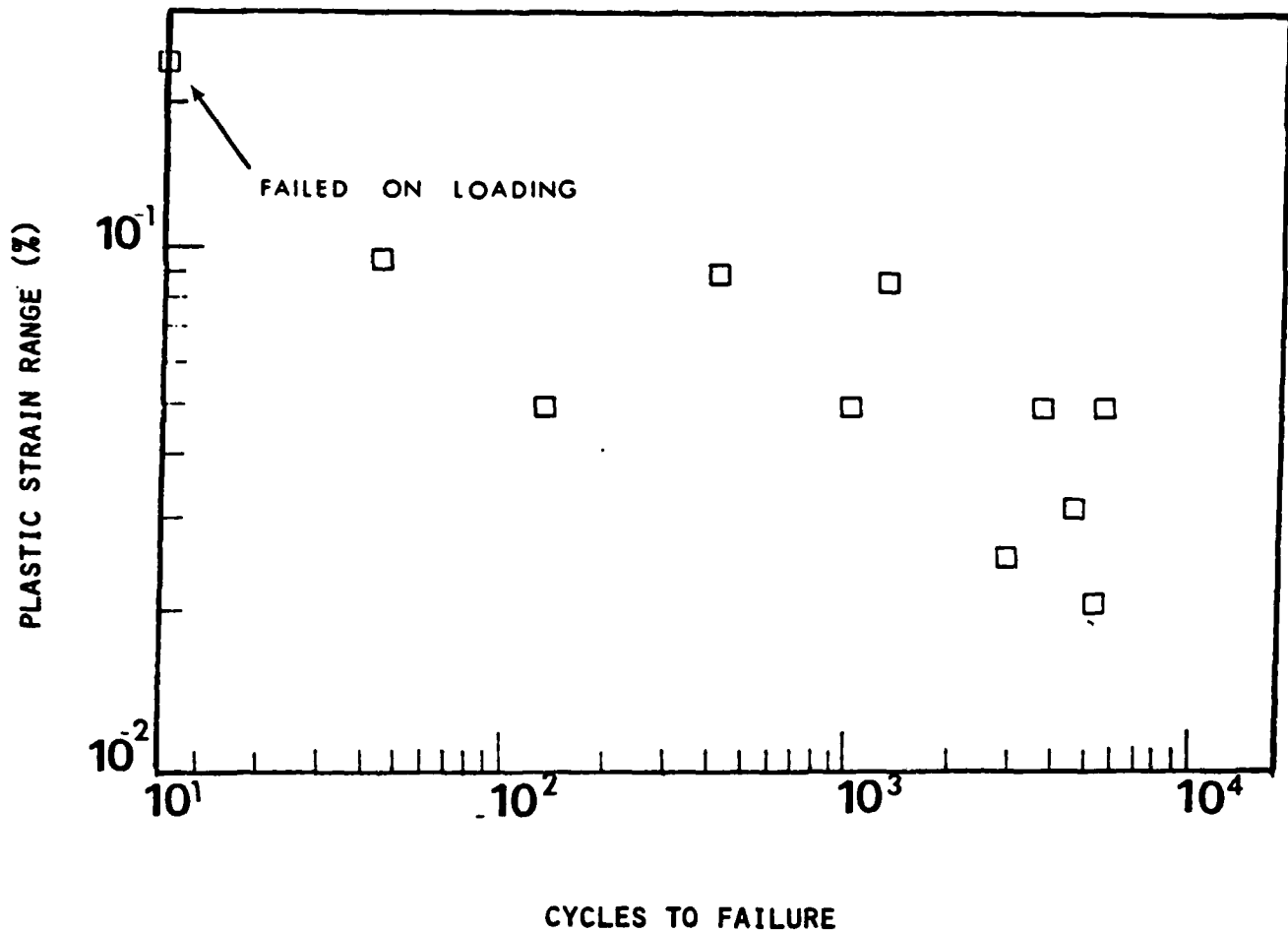


Figure 1. Room temperature LCF behavior of Rene' 80.  
All tests done at 0.55 Hz.

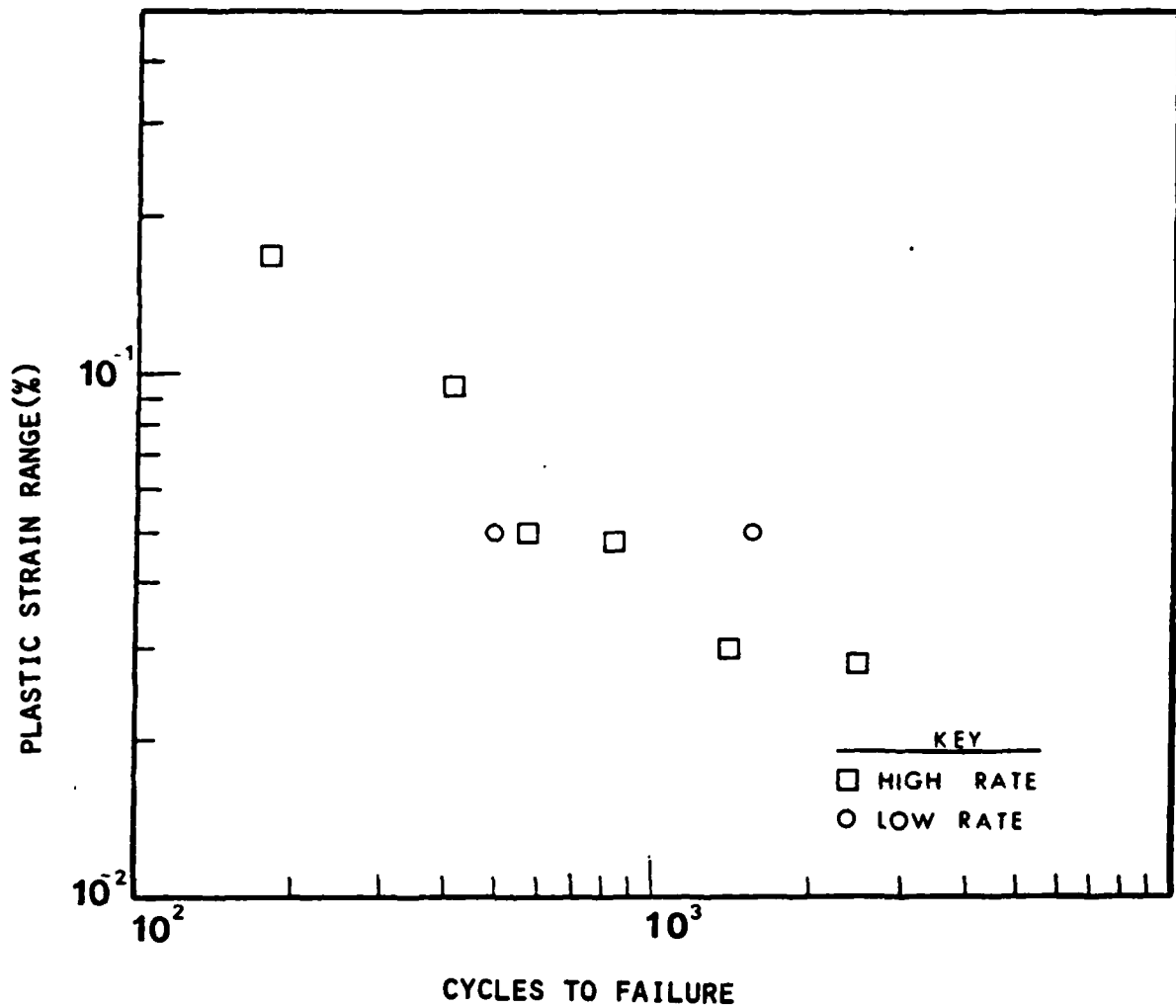
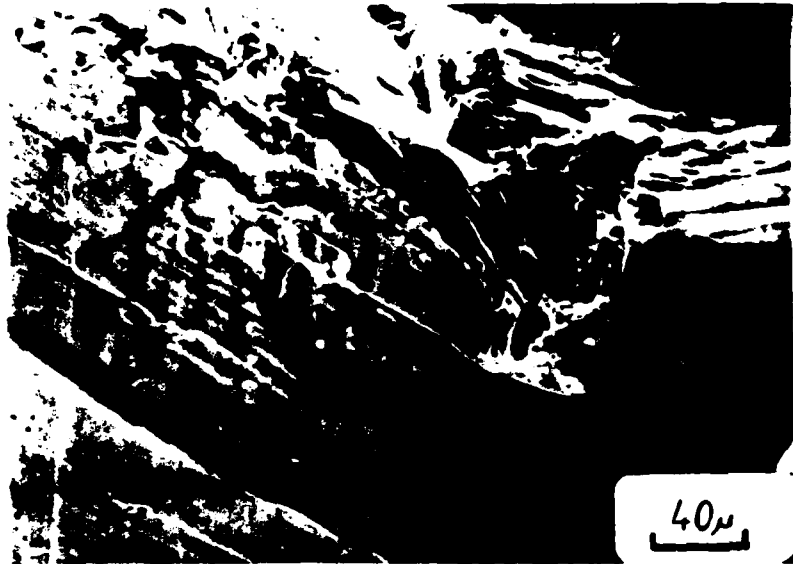


Figure 2. LCF behavior of Rene' 80 at 800°F.





(b)

Figure 3. SEM micrograph of initiation region in LCF specimen RTV2B. Note the crack formation along slip bands.  
-  $\Delta\epsilon_p=0.05\%$ ,  $N_f=3593$ ,  $T=700^\circ\text{F}$ .

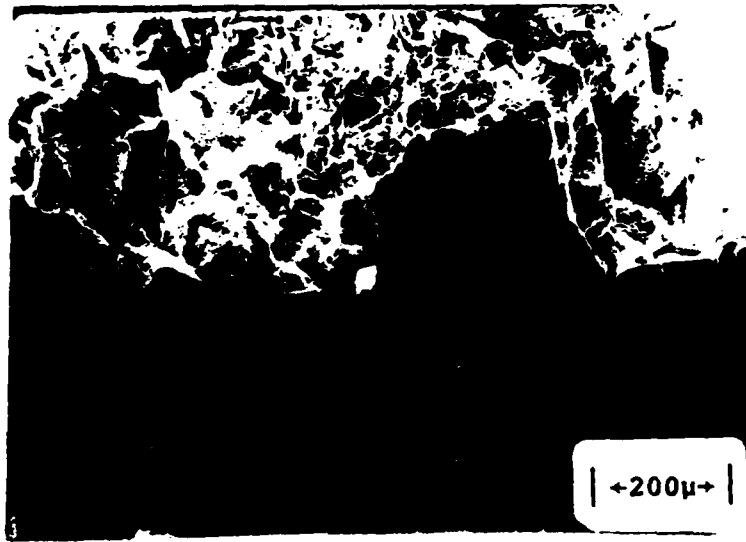


Figure 4. SEM micrograph of initiation region in LCF specimen RTV2A. Note the intergranular cracking which was associated with a reduced fatigue life.  $\Delta\epsilon_p=0.05\%$ ,  $N_f=995$ ,  $T=70^{\circ}\text{F}$ .

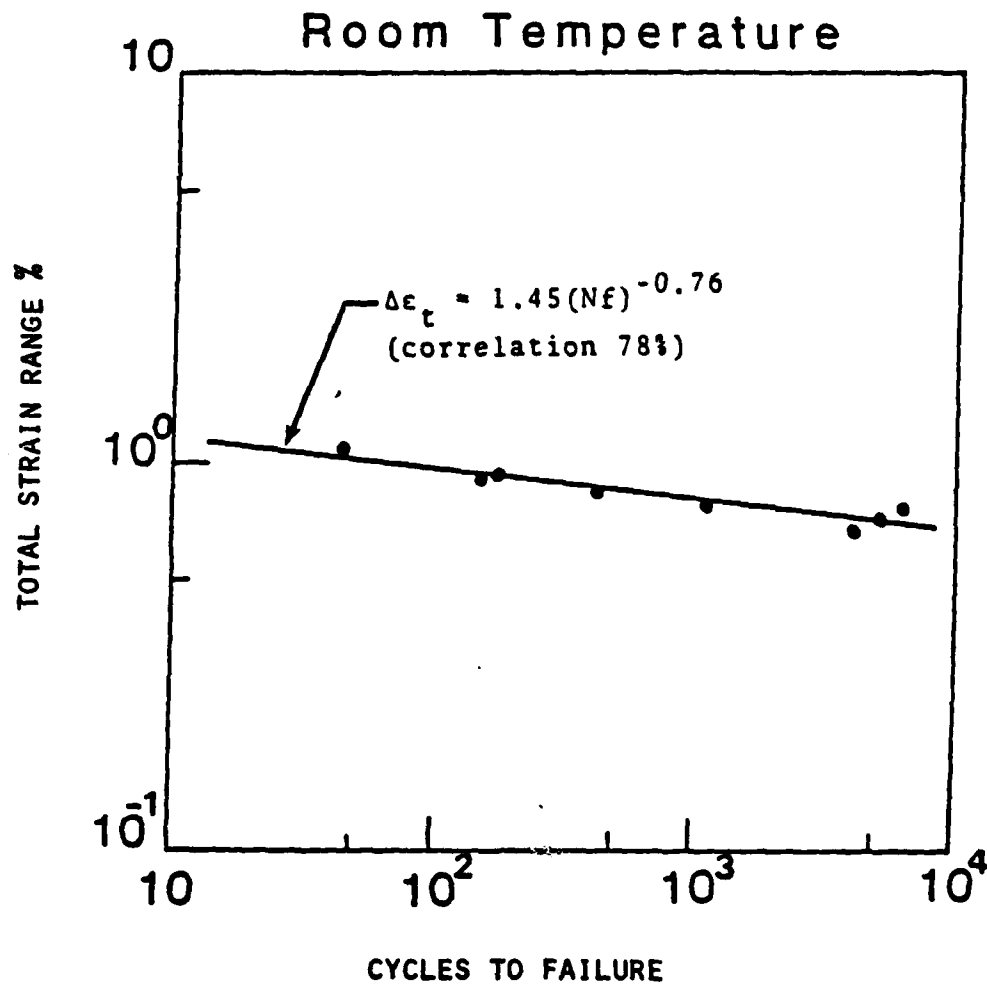


Figure 5. Plot of total strain range  $\Delta \epsilon_t$  versus number of cycles to failure  $N_f$  for RT LCF specimens. Note that the correlation is better than the one obtained from the Coffin-Manson curve in Fig. 1.

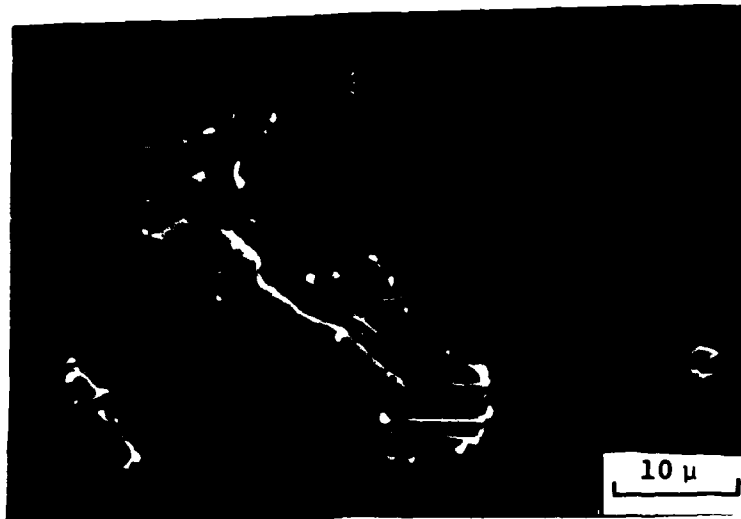


Figure 6. Crack initiation at the surface of a 75°F LCF specimen showing a cracked Ti carbide. Note that the cracks did not extend into the matrix nor was there any interfacial decohesion. Specimen 2H:  $\Delta\epsilon_p=0.055\%$ ,  $\Delta\epsilon_t=0.90\%$ ,  $N_f=372$ .

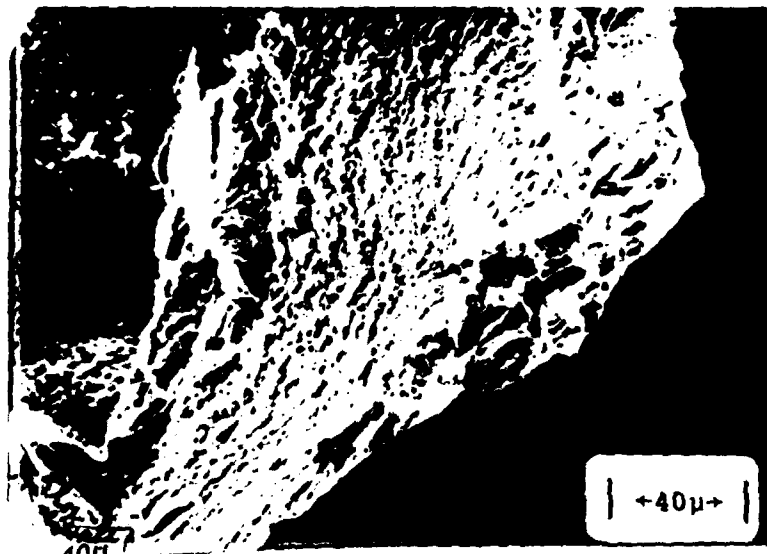


Figure 7. SEM micrograph of initiation region in LCF specimen 14LV2. Note initiation in grain boundary around Ti carbides.  $\Delta\epsilon_p=0.057\%$ .  $N_f=210$ ,  $T=1400^\circ\text{F}$ .

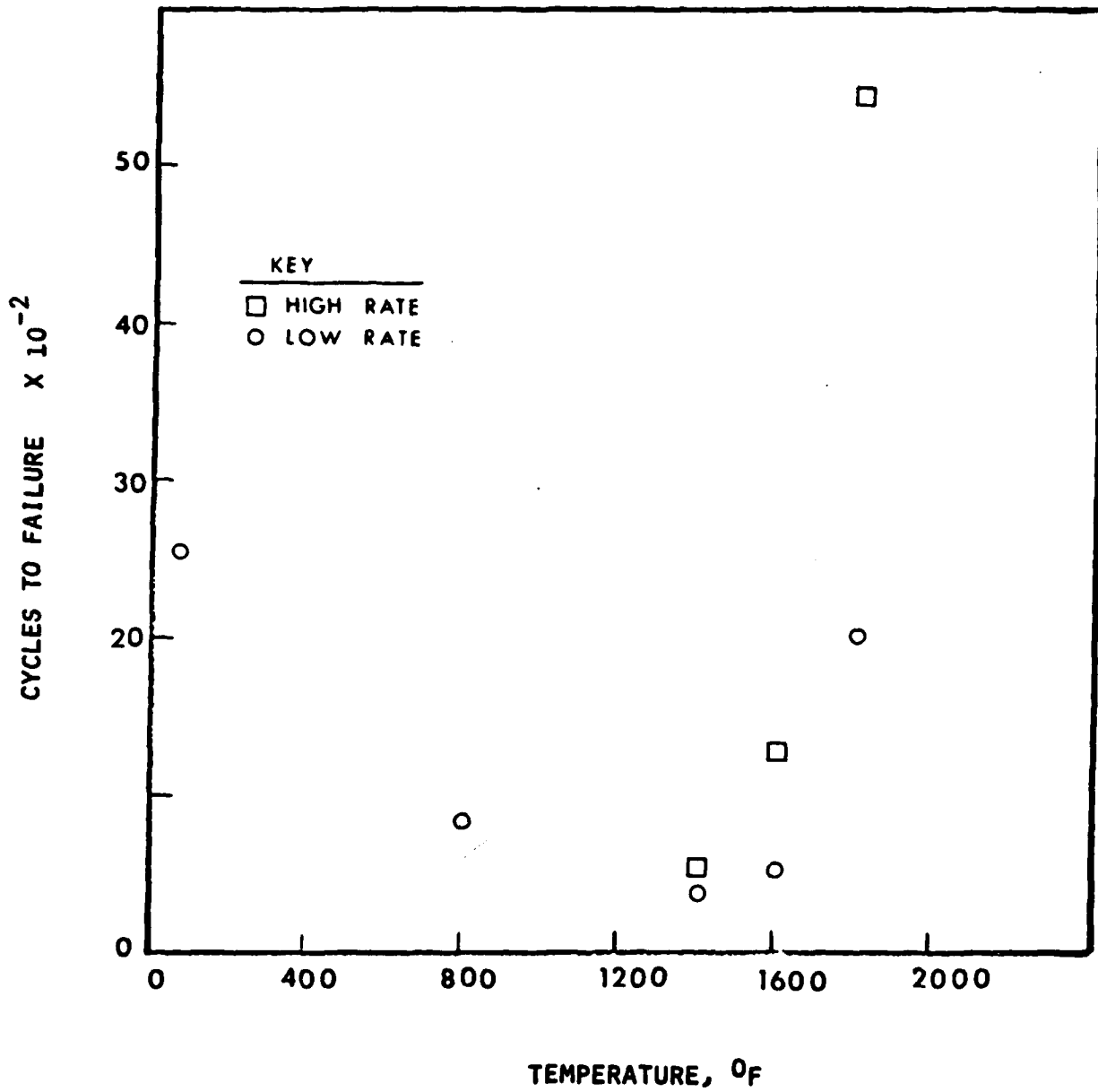


Figure 8. Effect of temperature on the LCF behavior of Rene' 80 at a plastic strain range of 0.1%.

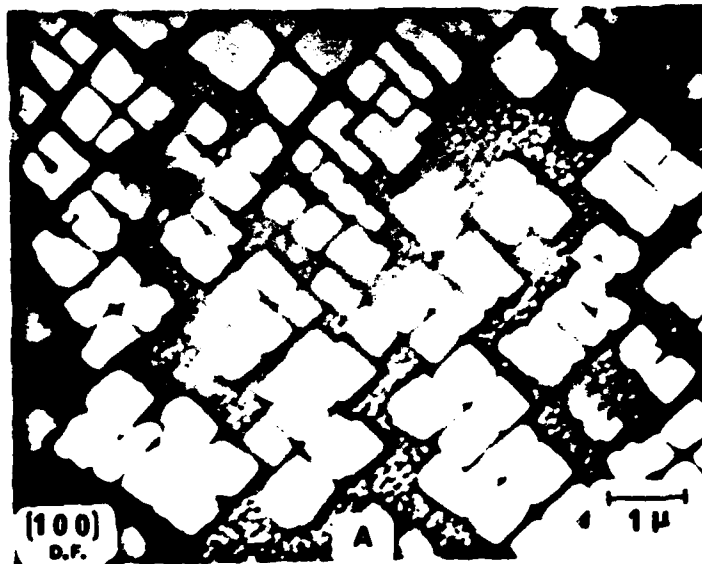
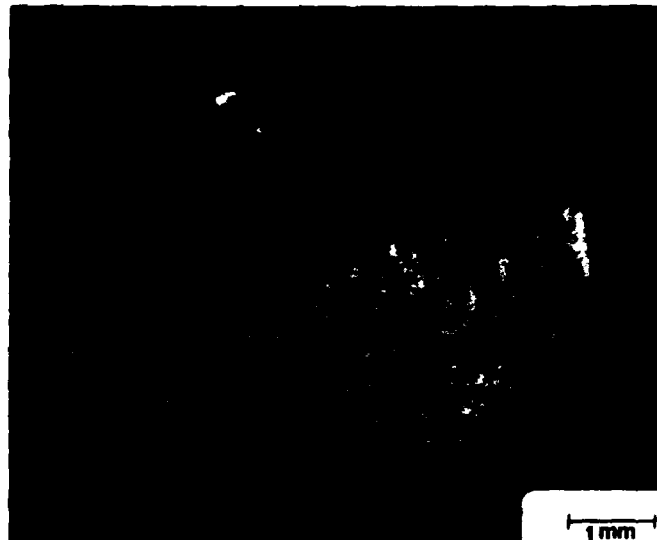
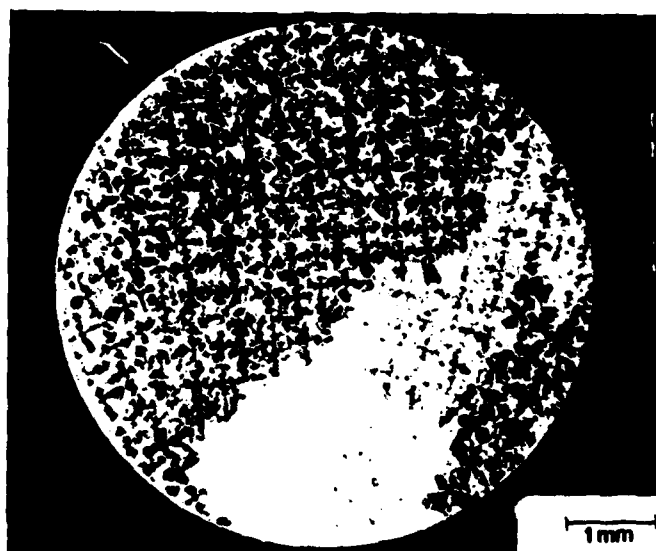


Figure 9. TEM micrograph of the as-heat treated  $\gamma'$  microstructure of standard composition DS Rene' 80 in (a) dark field and (b) bright field.



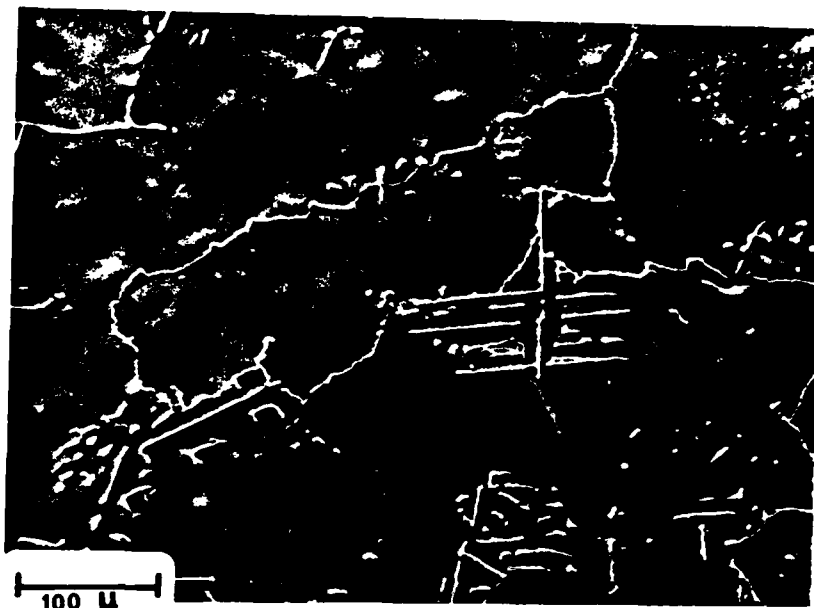
(a)



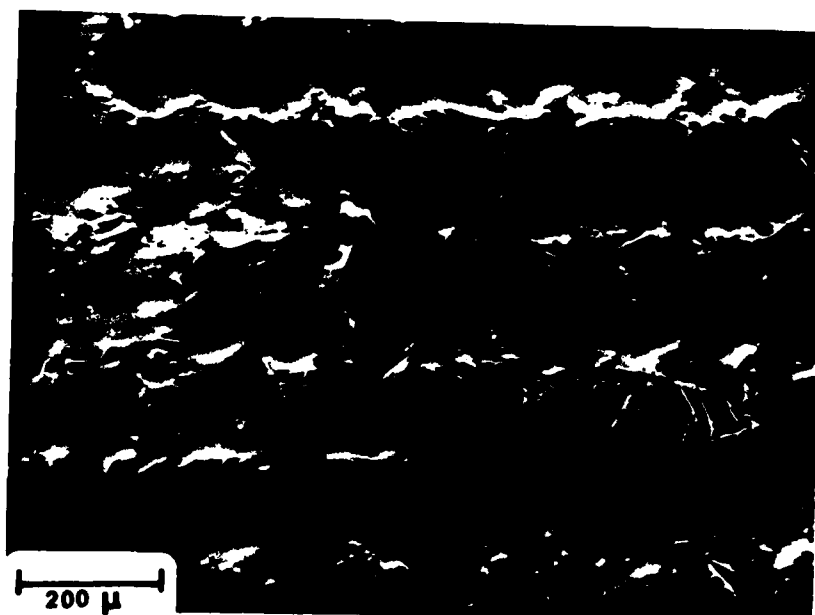
(b)

Figure 10. Typical grain configuration in the transverse cross section of longitudinal LCF samples taken from (a) bottom and (b) top of a DS Rene' 80 bar.





(a)



(b)

Figure 11. Typical carbide structures in DS Rene' 80.  
(a) is transverse to the growth direction of the bar and (b) is parallel to the growth direction.

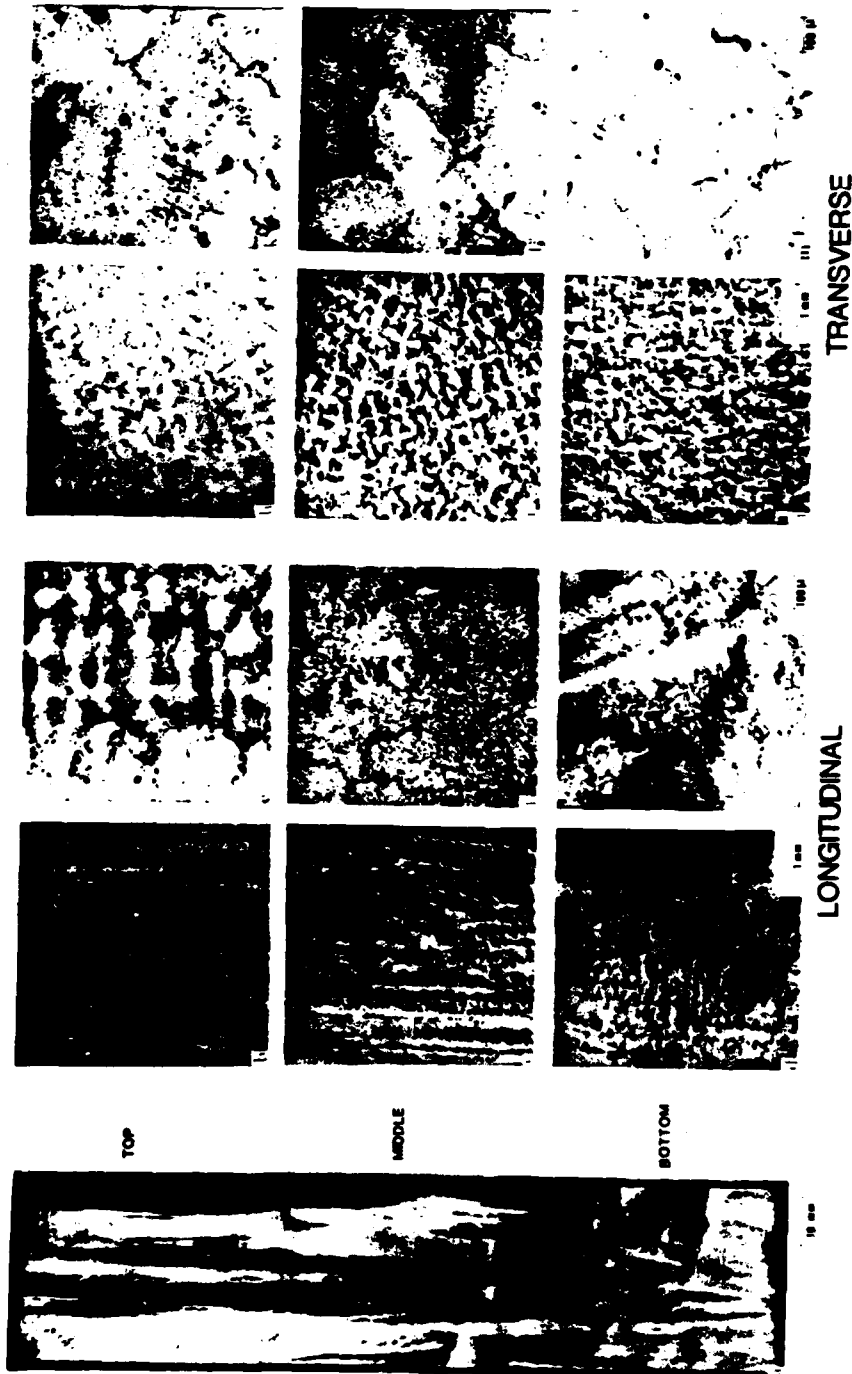


Figure 12. Variation in carbide morphology and grain size as a function of the distance from the chilled end in directionally solidified Rene' 80.

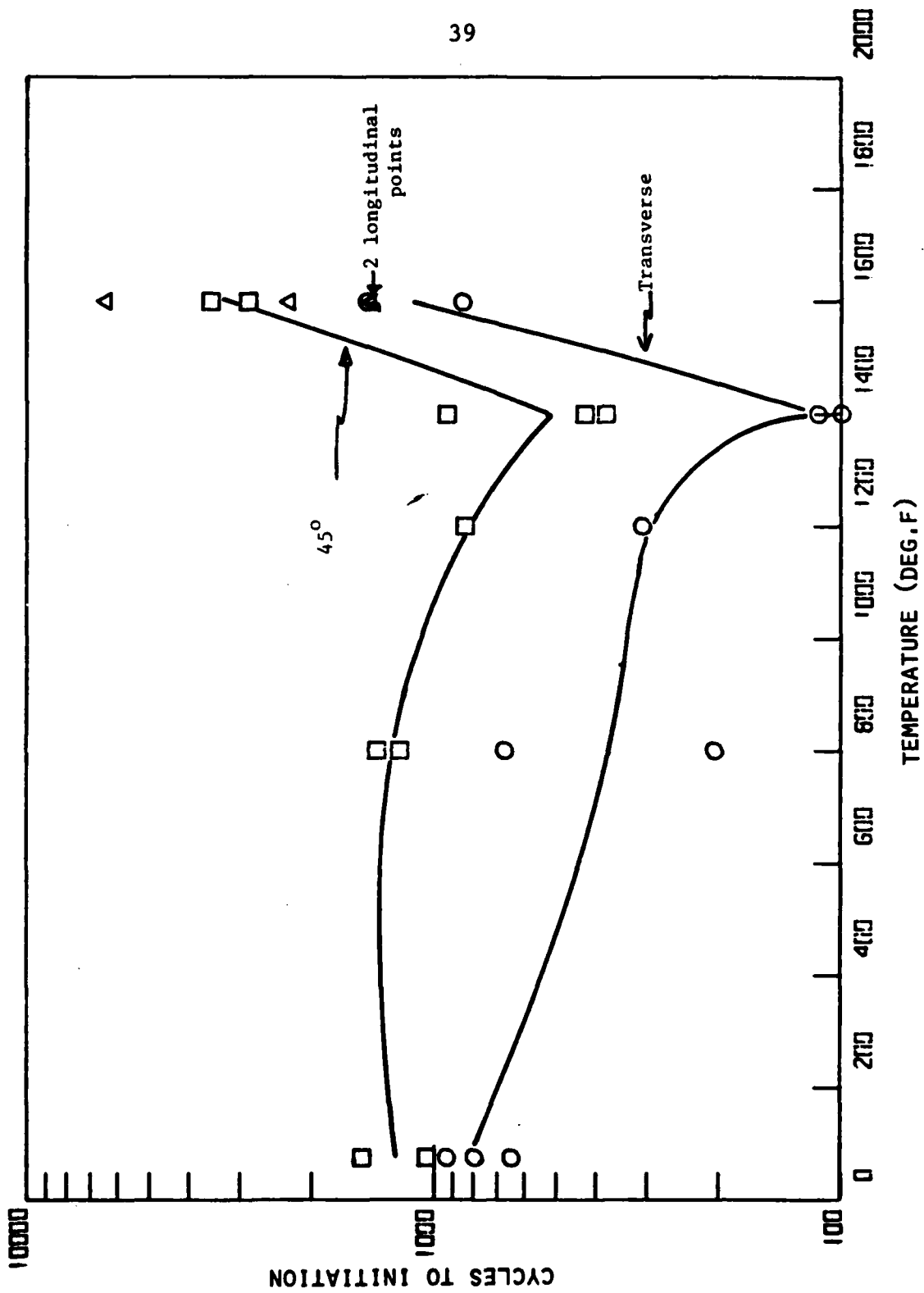


Figure 13. Cycles to initiation vs. temperature for standard Rene' 80. 0.1% plastic strain range. Triangles are longitudinal, squares are 45°, and circles are transverse.

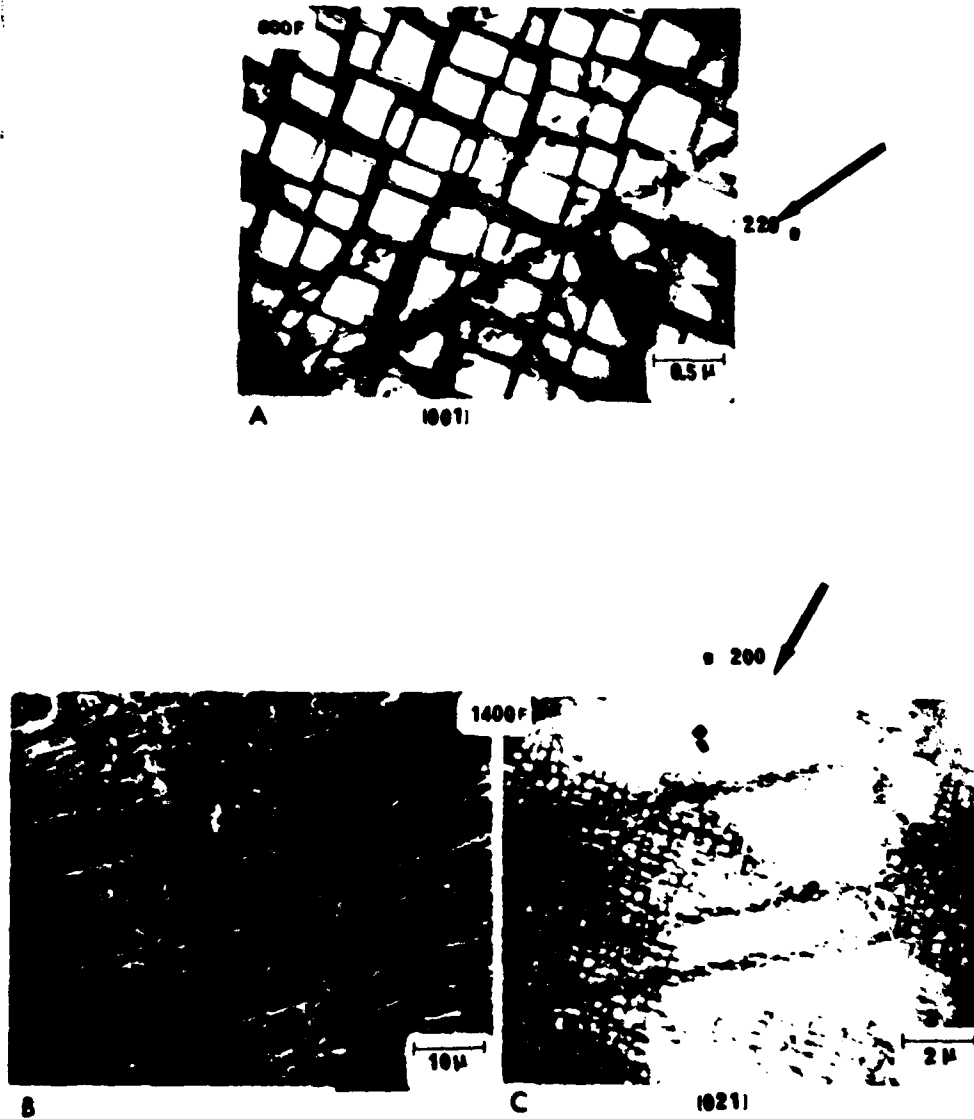


Figure 14. LCF specimens tested up to 1400°F showing slip bands.  
 (A) TEM, 800F,  $\Delta\epsilon_p=0.1\%$ ,  $N_1=205$ , (B) SEM 1400F,  
 $\Delta\epsilon_p=0.1\%$ ,  $N_1=930$ . Note oxidized surface  
 (C) TEM 1400F,  $\Delta\epsilon_p=0.1\%$ ,  $N_1=114$ .



Figure 15. Shearing of  $\gamma'$  as a result of planar slip in a specimen tested at 75F,  $\Delta\epsilon_p=0.25\%$ ,  $N_1=1125$  (494-B-7).

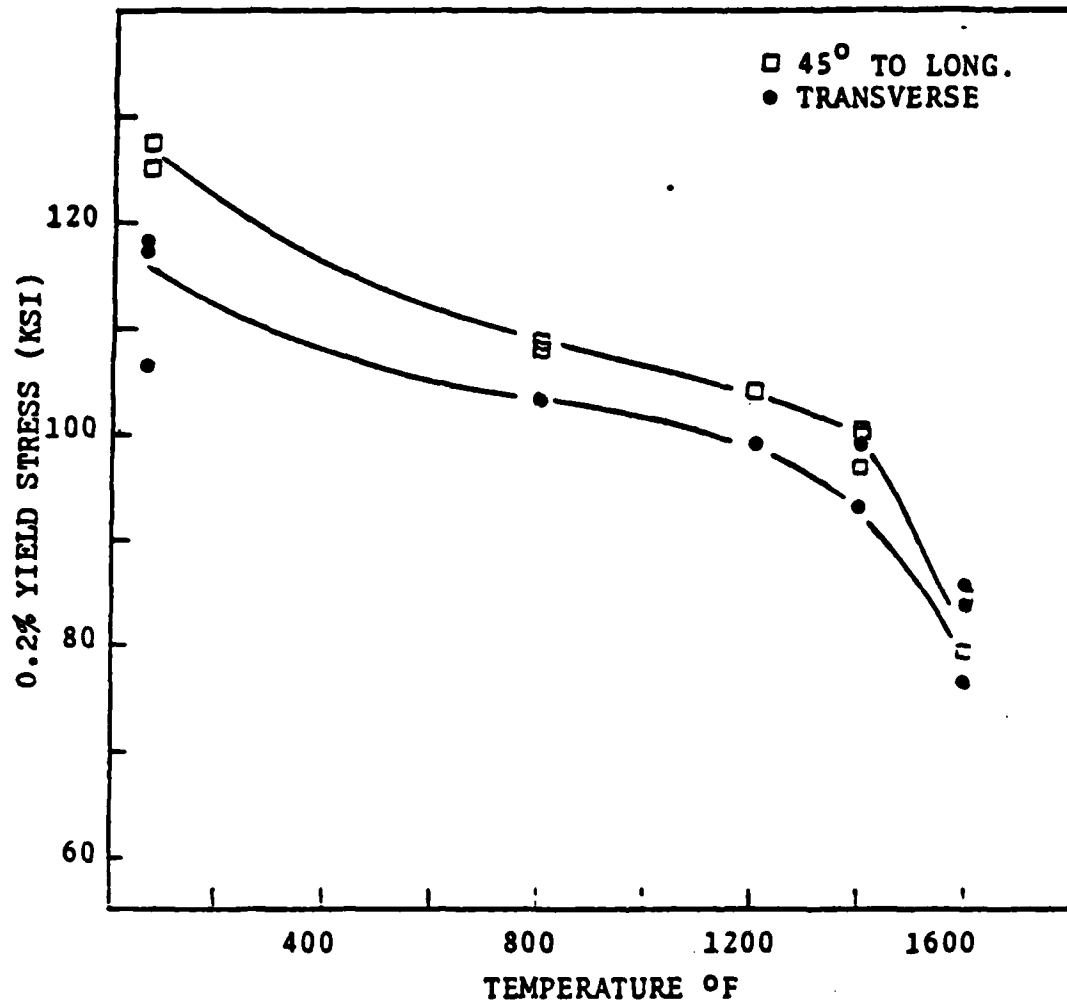


Figure 16. 0.2% yield stress of DS Rene' 80 as a function of temperature. The yield stress was measured using the first loading cycle of the LCF tests.

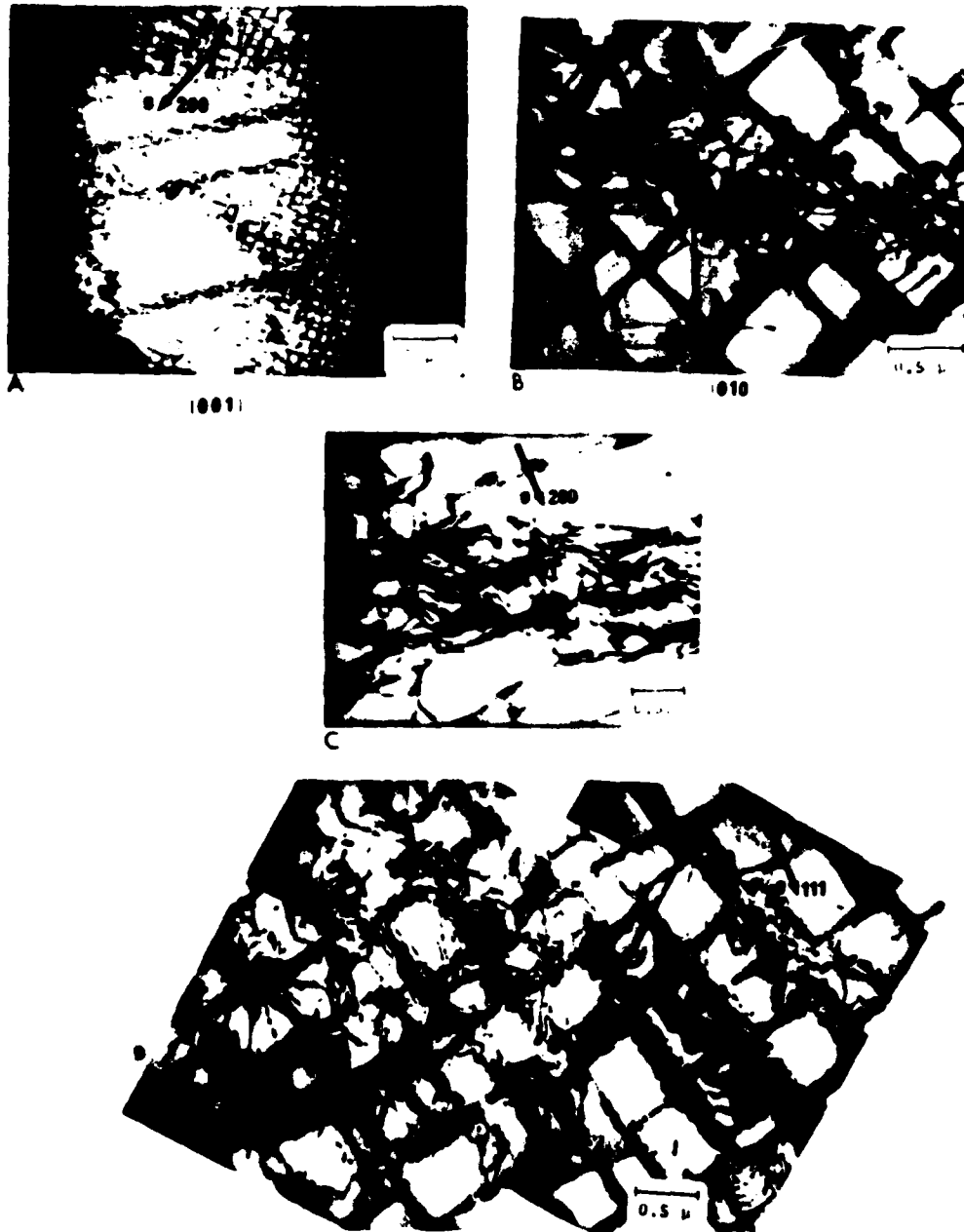
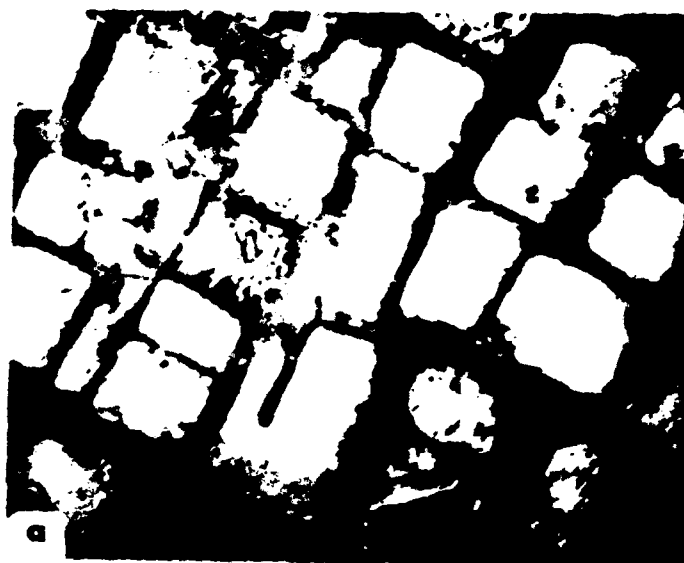
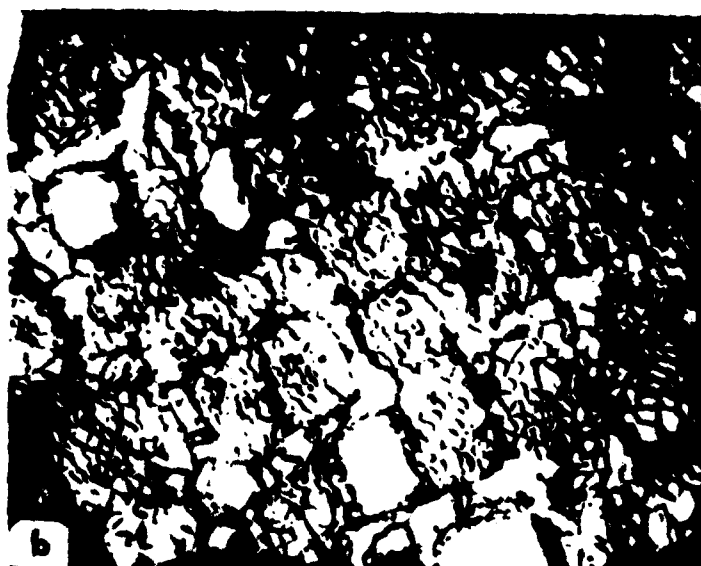


Figure 17. Planar slip bands formed during LCF testing at 1400°F (A and B). The dislocations are found preferentially in the regions between the large precipitates forming walls around the  $\gamma'$  (C and D). This temperature marked a transition from planar to homogeneous slip.  $\Delta\epsilon_p=0.1\%$ ,  $N_f=114$ .



0.5  $\mu$



$\{200\}$

Figure 18. Typical structure as seen by TEM after LCF at 1600°F. It can be seen that the large cuboidal precipitates have grown while the small ones have disappeared in the dark field micrograph (a), while a network of interfacial dislocations is visible in (b).  $\Delta\epsilon_p = 0.25\%$ ,  $N_1 = 260$ .



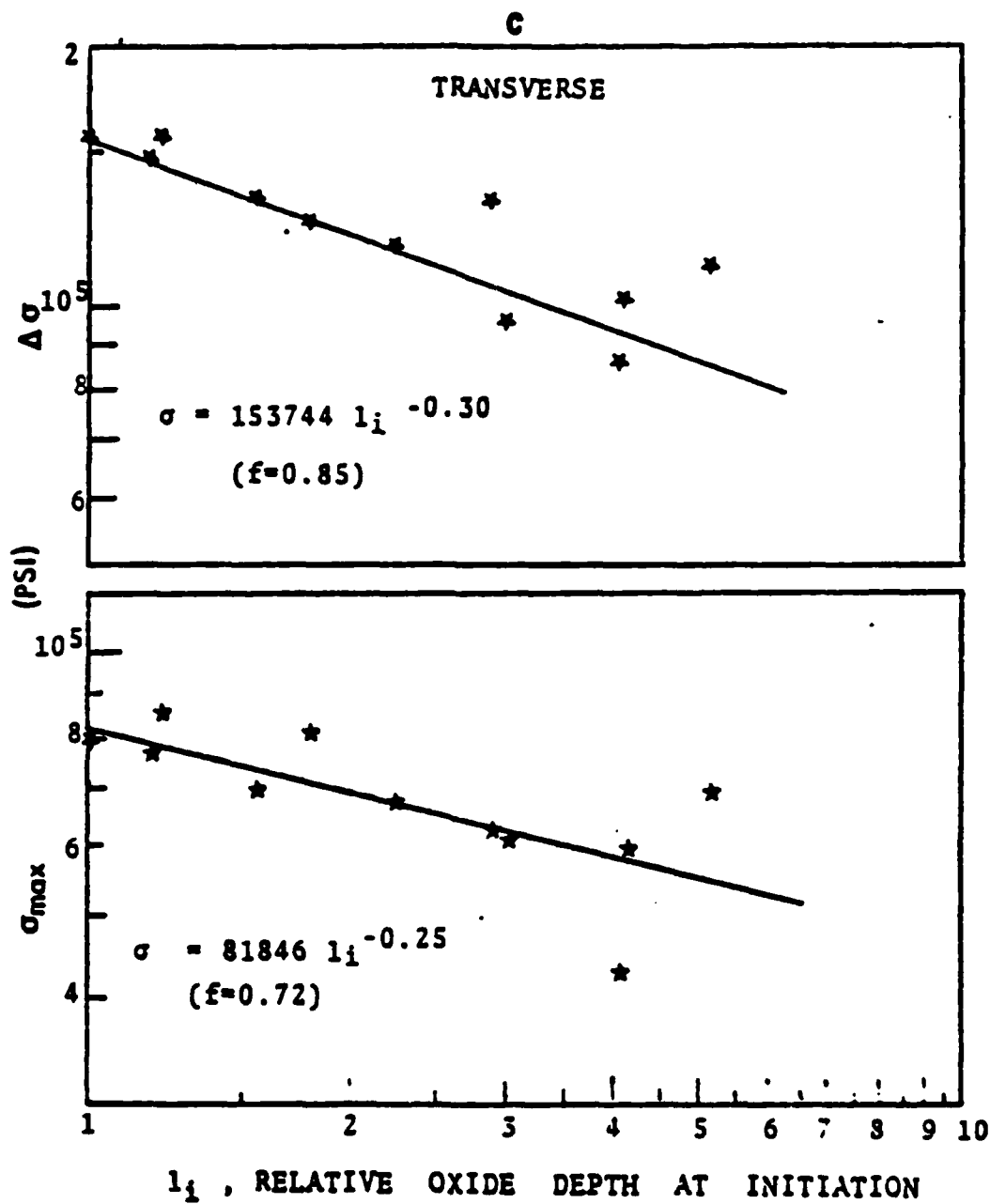


Figure 19. Stress range ( $\Delta\sigma$ ) or maximum stress ( $\sigma_{max}$ ) as a function of the relative oxide depth at initiation for DS Rene' 80 in the transverse orientation. All testing was done at 1600°F.

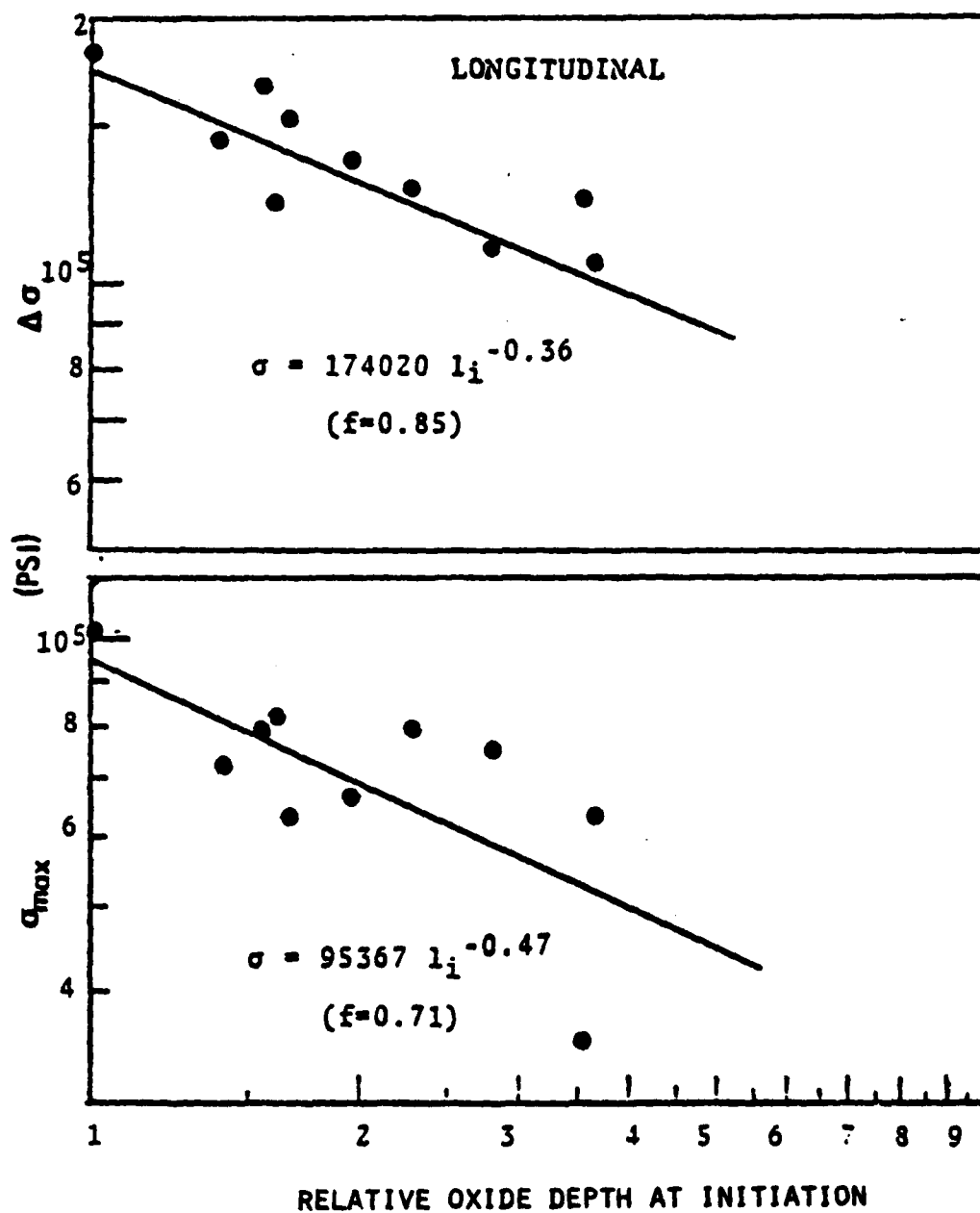


Figure 20. Same as Fig. 19 except for longitudinal specimens.

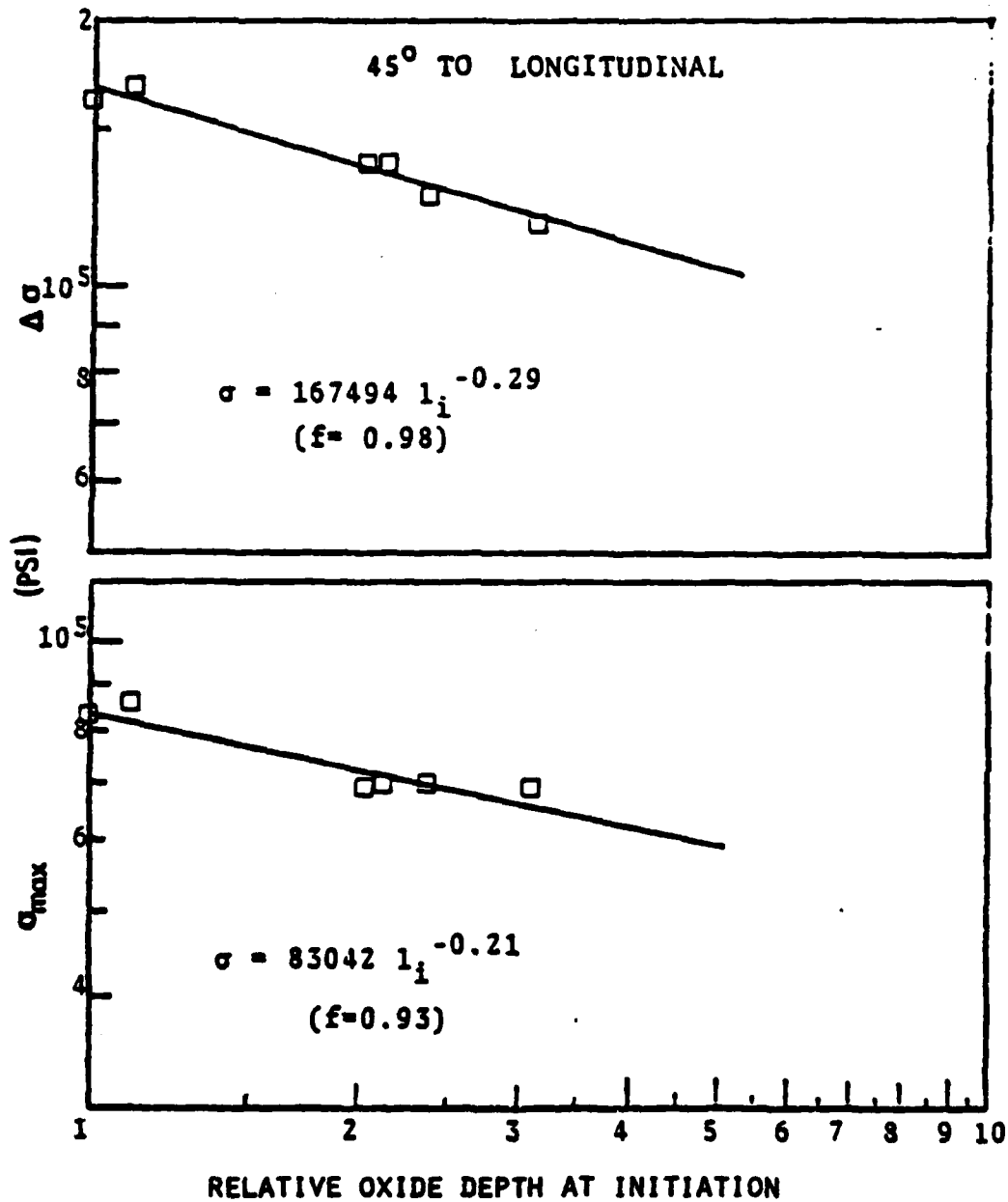


Figure 21. Stress range ( $\Delta\sigma$ ) or maximum stress ( $\sigma_{max}$ ) as a function of the relative oxide depth at initiation for DS Rene' 80 at 45° to longitudinal direction. All testing was done at 1600°F.

**$N_1$  - CYCLES TO INITIATION**

**Figure 22. Coffin-Manson plot for transverse specimens tested at 1600F. The position in the ingot is noted.**

**Figure 23. Coffin-Manson plot for longitudinal specimens tested at 1600F. The position in the ingot is noted.**

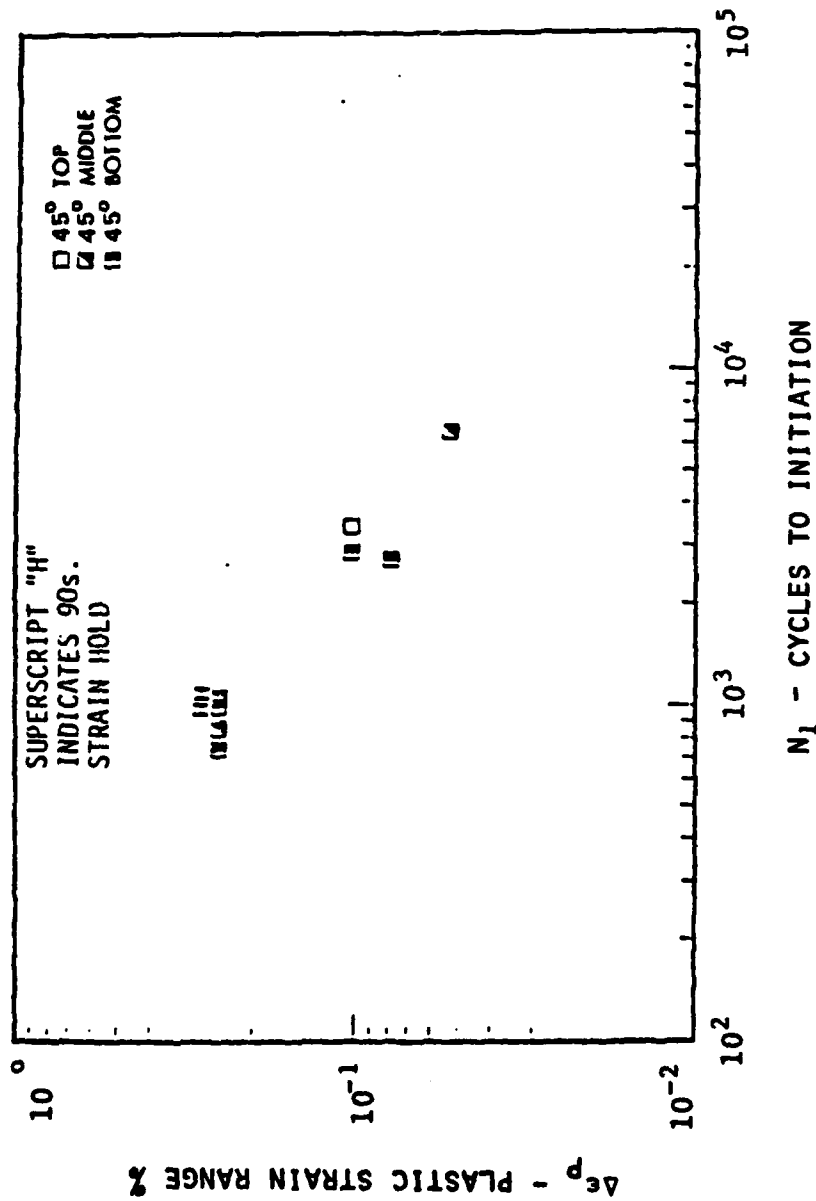


Figure 24. Coffin-Manson plot for 450 specimens tested at 1600F. The position in the ingot is noted.

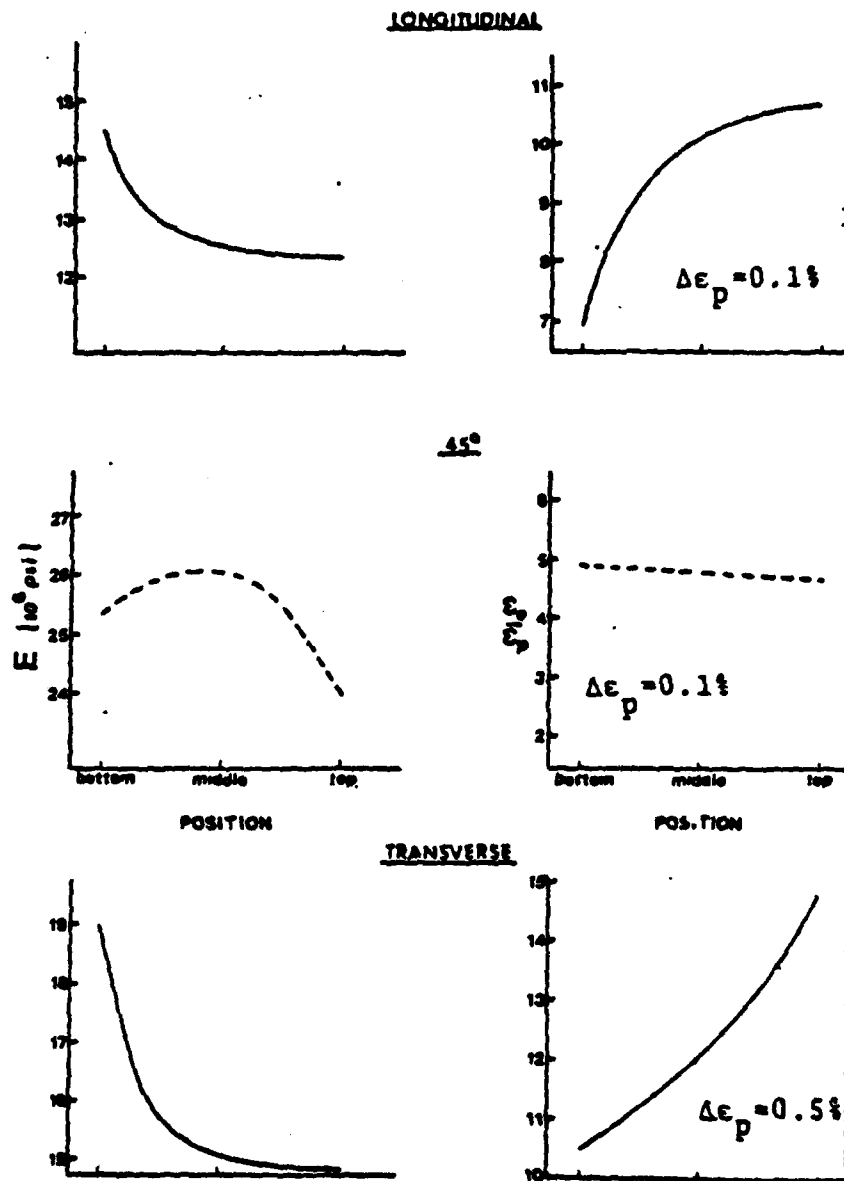


Figure 25. Effect of location on modulus and the elastic strain associated with a given plastic strain. This behavior appears to be related to the misorientation from the [001] growth axis.

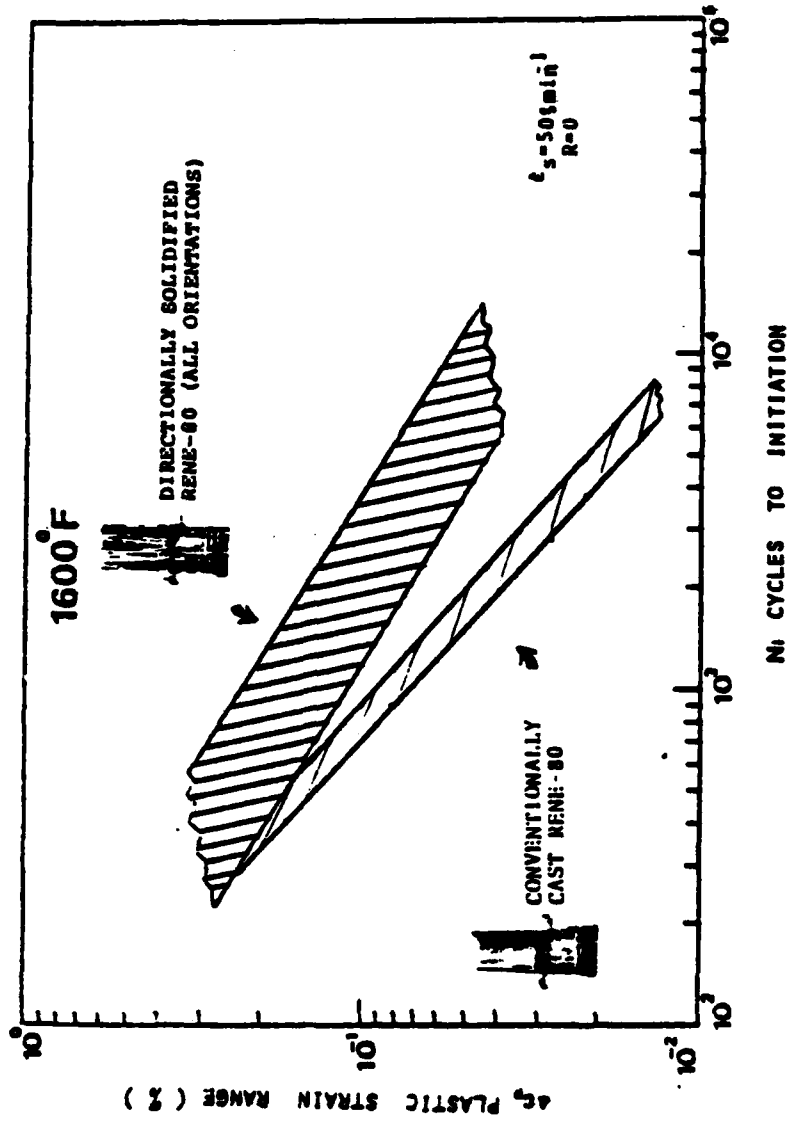
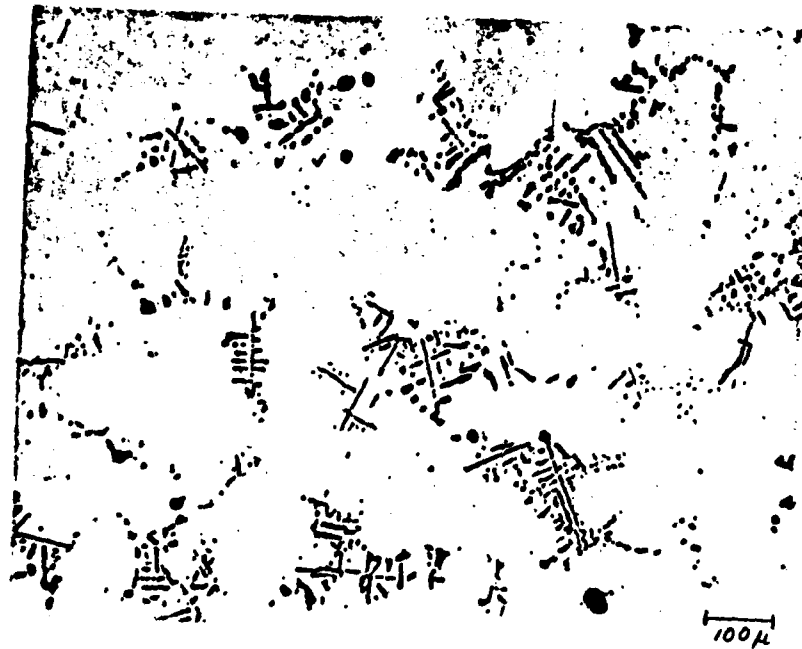


Figure 26. LCF behavior of DS Rene' 80 at 1600°F.





(a)



(b)

Figure 27. Typical microstructure of (a) DS Rene' 80 and (b) Hf modified DS Rene' 80.

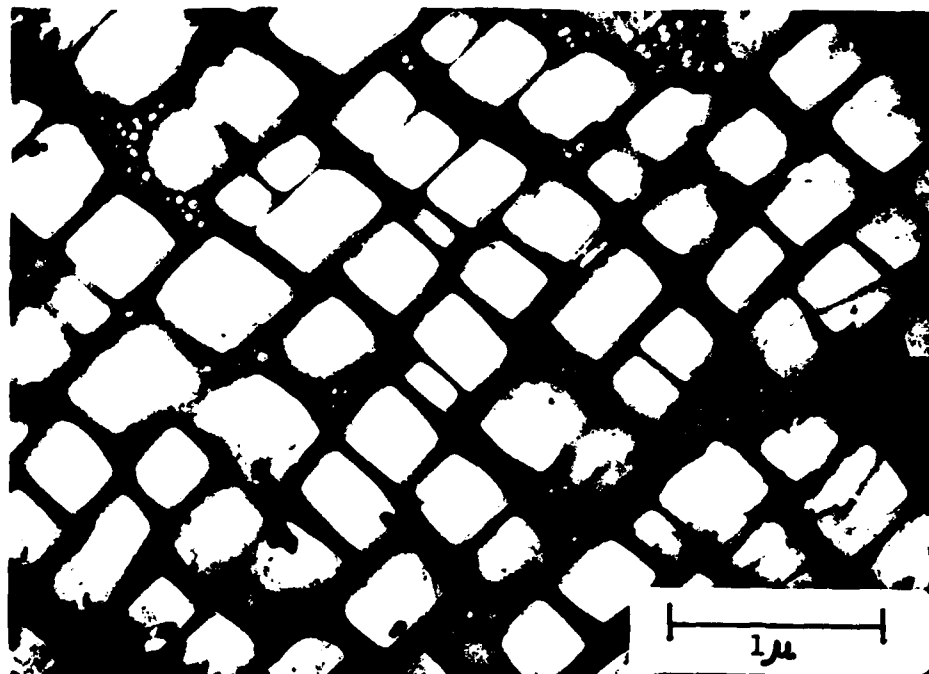
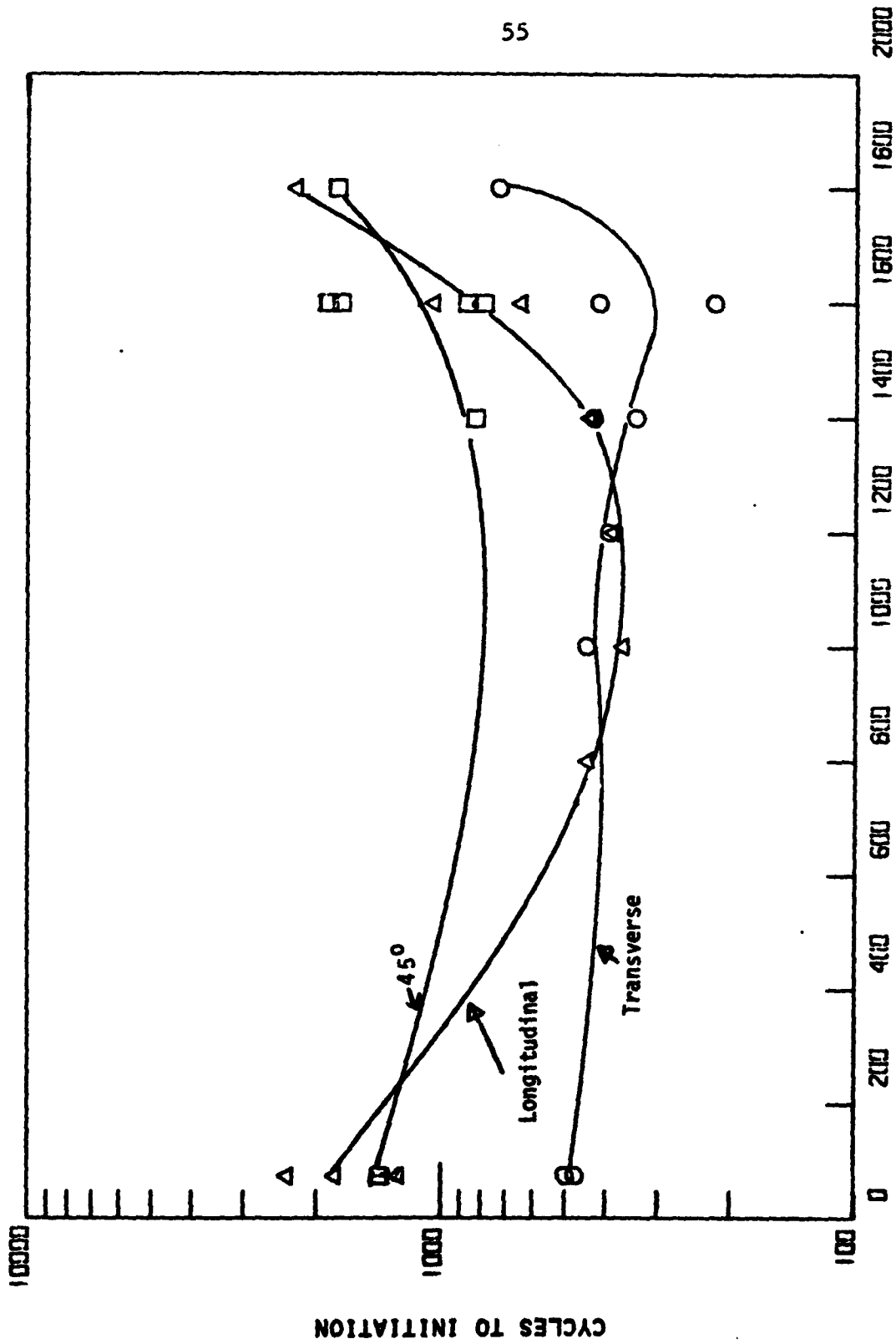


Figure 28. Dark field TEM micrograph of the  $\gamma^1$  microstructure of Hf modified DS Rene' 80.



TEMPERATURE (DEG.F)

Figure 29. Cycles to initiation vs. temperature for Hf modified DS Rene' 80 at 0.1% plastic strain range. Triangles are longitudinal, squares are 45° and circles are transverse.

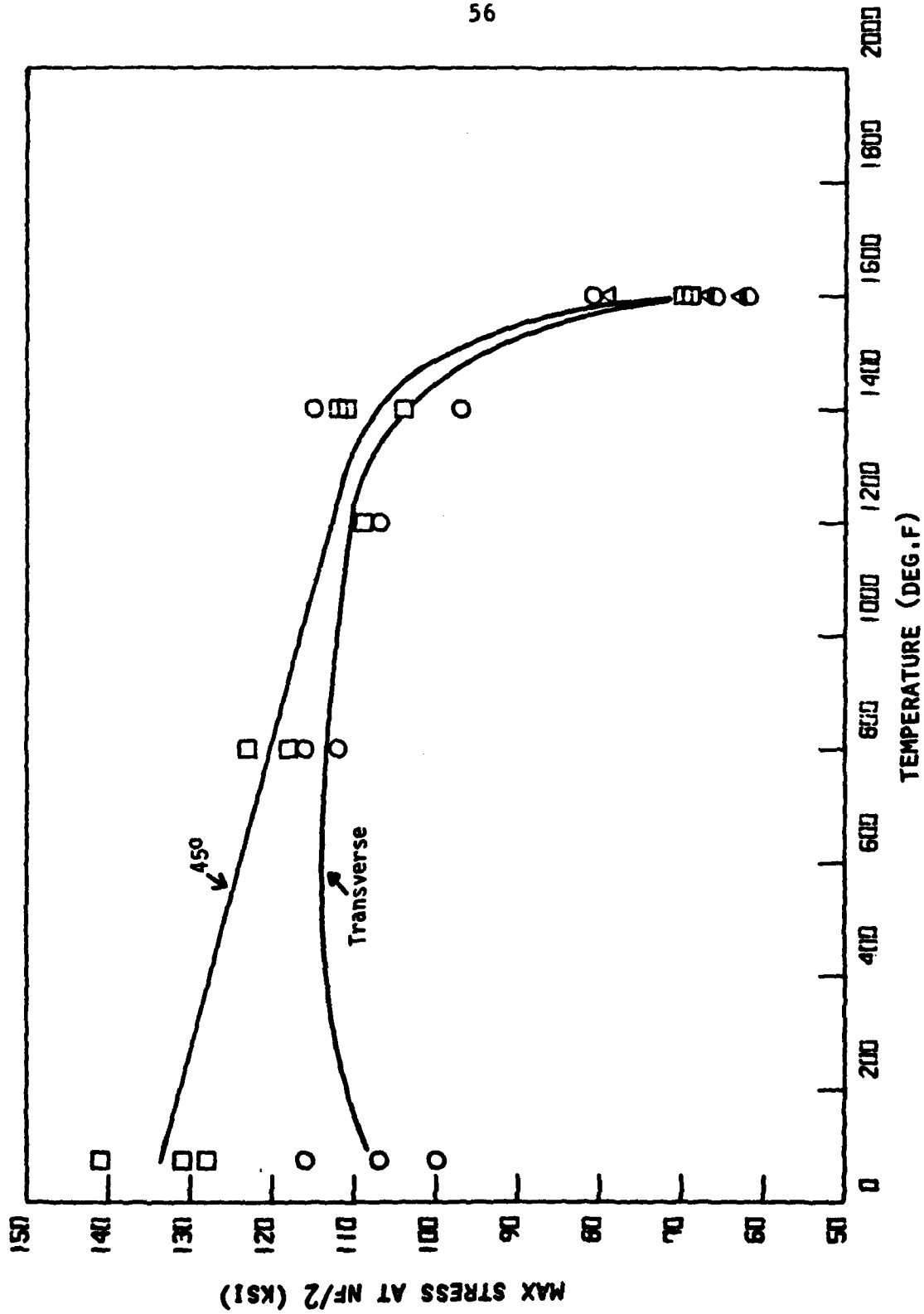


Figure 30. Max stress at NF/2 vs. temperature for standard Rene' 80 at 0.1% plastic strain range. Triangles are longitudinal, squares are 45° and circles are transverse.

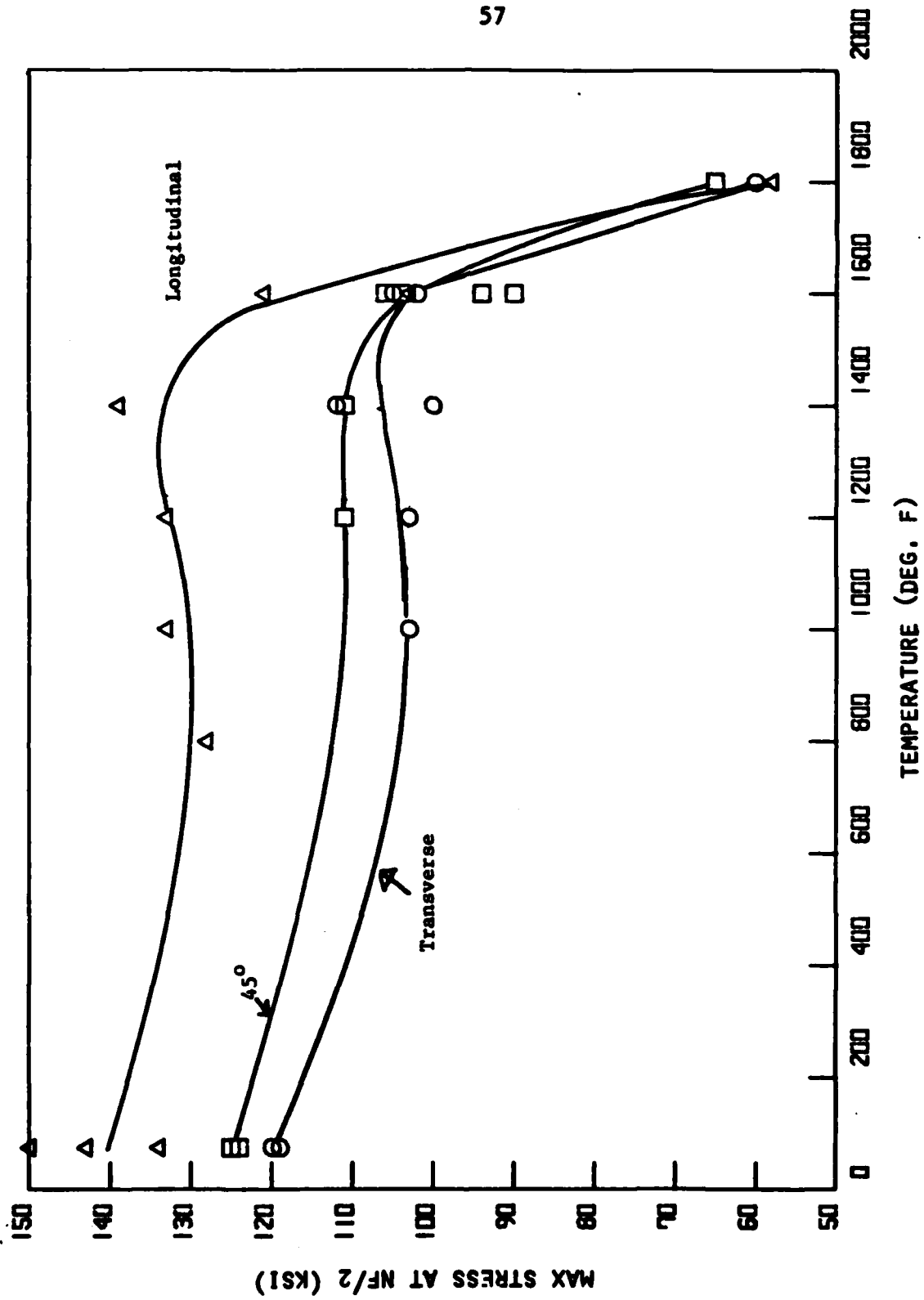


Figure 31. Max stress at NF/2 vs. temperature for Hf modified Rene' 80. 0.1% plastic strain range. Triangles are longitudinal, squares are 45° and circles are transverse.

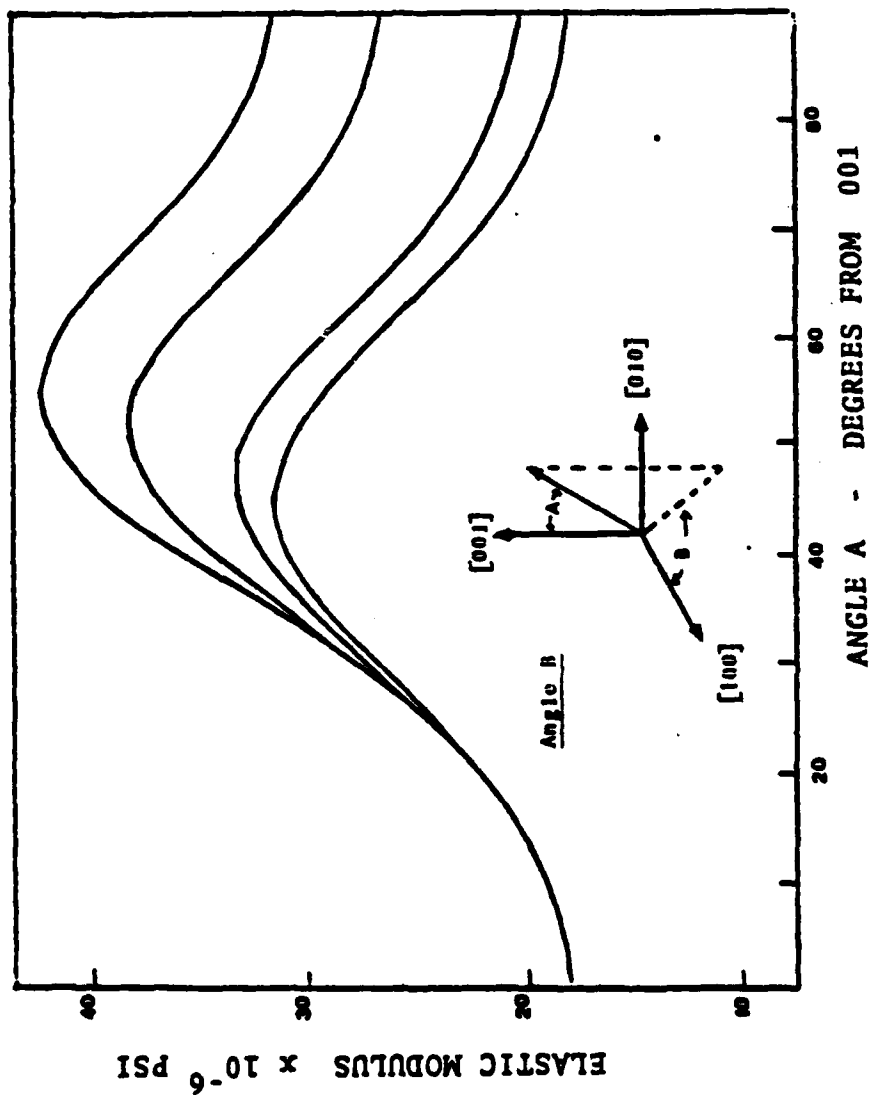


Figure 32. Estimated elastic modulus variation in single crystal nickel.

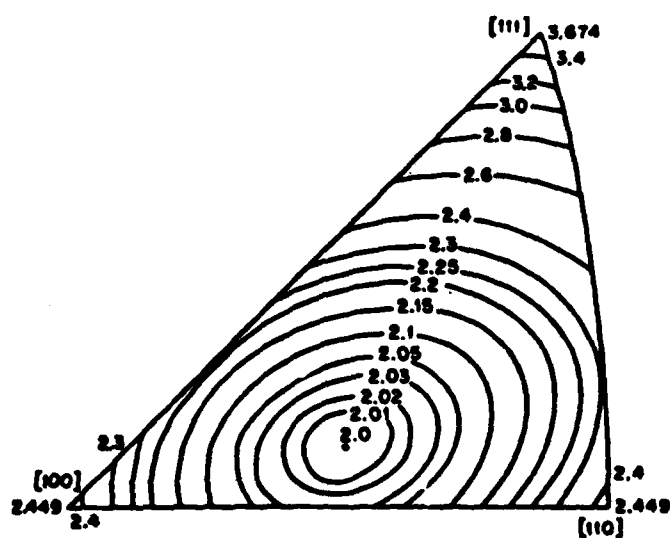


Figure 33. Contours of constant inverse Schmid factor  $m^{-1} = (\cos \lambda \cos \phi)^{-1}$ , for uniaxial tension based on  $\{111\}\langle 110 \rangle$  slip.

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